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Creep-Fatigue Fracture: Analysis of Internal Damage

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Preface

Energy-related high-temperature equipment such as thermal power generation facilities is indispensable to human life. In order to utilize them safely and effectively, it is important to elucidate the mechanism of high-temperature creep-fatigue of materials. High-temperature creep-fatigue fracture of heat-resistant steel can be categorized into two types: “surface cracking type fracture” and “inner cracking type fracture.” The former fracture is caused by the growth and coalescence of many small cracks initiated on the surface of the material, while the latter fracture is caused by the growth and coalescence of multiple cavities and small cracks initiated inside the material. Both types of fractures follow an extremely complicated process, whereby small defects, such as cavities and small cracks, are initiated randomly in terms of both time and location and subsequently grow with coalescence. Therefore, even if creep-fatigue tests are performed on specimens made of the same material under the same conditions, there might be differences in the accumulation of creep-fatigue damage from specimen to specimen, resulting in variations in fatigue lives of specimens. Although predicting the initiation of cracks remains challenging, continuum fracture mechanics, including creep J-integral and creep J-integral range, have been used to predict crack growth behavior. However, these parameters of fracture mechanics cannot be applied to cracks whose size is similar to or smaller than the microstructure such as grains, because the material surrounding the crack cannot be considered as a homogeneous body. Thus, accurately predicting the behavior of small crack initiation and growth, which account for the majority of fatigue lives, remains a significant difficult problem. In the case of inner cracking type fracture, small cracks are often initiated by the coalescence of cavities that initiate and grow on grain boundaries. Analytical investigations of creep cavities have also been conducted using dislocation theory and diffusion theory, and observation results have been reported for heat-resistant steels. Generally, cavities are observed on the surface of the material or on cut surfaces inside the material, so it is difficult to accurately grasp the true size distribution inside the material. In the past, various parameters such as the A-parameter have been proposed for the purpose of evaluating damage caused by cavities, and have been used to evaluate the remaining life of actual machines. Regarding the parameters used for damage evaluation, it is desirable to use appropriate parameters

after clarifying the actual cavity distribution and the physical meaning of the damage parameters. However, this subject has not been rigorously discussed.

Based on the above-mentioned background, this book provides a detailed description of creep-fatigue fracture of austenitic stainless steel SUS304, which is a typical heat-resistant steel used in power generation equipment and chemical plants, and steam turbine rotor material 1Cr-1Mo-1/4V forged steel. In particular, for inner cracking type fractures in SUS304, the entire fracture processes from creep cavities to small cracks and macrocracks were observed in detail, and three-dimensional numerical simulations were performed based on the obtained results. The three-dimensional numerical simulations successfully visualized the initiation and growth behavior of small cracks inside the material, which cannot be seen by ordinary experimental observation. Furthermore, to contribute to extending the life of aged materials, we proposed a method of extending the life of damaged materials with numerous small inner cracks by applying a few number of cyclic loads with a special strain waveform to the material and were able to confirm its effectiveness through experiments. Although this book presents results for two representative types of heat-resistant steels, the obtained results are not limited to these materials, but can also be applied to creep-fatigue fractures of other heat-resistant steels that exhibit similar fractures.

Finally, we would like to express my sincere gratitude to Profs. Ryuichi Otani and Takayuki Kitamura, Professors Emeritus at Kyoto University, for their advice and guidance in conducting the series of research outlined in this book.

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Chapter 1

Introduction



Abstract This chapter first presents the engineering and scientific background on high-temperature creep-fatigue. Secondly, research and problems regarding conventional high-temperature creep-fatigue are pointed out. Finally, the three objectives and significance of this book are described.

1.1 Engineering and Academic Background

When metallic materials are exposed to high temperatures, thermal energy increases the movement of atoms, which not only reduces their strength but also causes various phenomena that differ from those at room temperature. One of the most important mechanical phenomena is creep. Creep is a nonlinear and time-dependent phenomenon and is the most important phenomenon in the deformation and fracture of metallic materials at high temperatures. The mechanism of creep and its dependence on temperature differs depending on the temperature level. Furthermore, it is an extremely complex phenomenon as it changes depending on loading conditions such as stress and strain rate even at the same temperature [1–3]. For example, even in the high-temperature creep temperature range, various fracture mechanisms appear depending on the temperature, stress, and strain rate such as voids due to transgranular deformation, grain boundary cracks due to grain boundary slip, and grain boundary cavities due to grain boundary diffusion creep [2–5]. Furthermore, the fracture mechanism and morphology of creep-fatigue, in which stress changes cyclically, are basically the same as those of static creep, but the addition of cyclic effects of stress or strain results in more complicated behavior [6].

In Japan, the number of aged components in energy-related high-temperature equipment and structures, such as thermal and nuclear power generation facilities that were constructed during the period of high economic growth, is rapidly increasing and reaching the end of their design lives one after another [7–9]. However, due to the difficulty in locating new equipment and the pursuit of economic efficiency, there is a growing trend to extend the life of aging equipment, and there is a need to improve the reliability of remaining life evaluation for existing high-temperature equipment with

higher accuracy. However, in Japan, damage tolerance design is not permitted for most power generation equipment. In other words, the ostensibly acceptable crack initiation is limited to practical and industrial crack initiation (cracks of approximately 0.1 mm in size) [6], the small crack initiation and growth addressed in this book correspond to the crack initiation processes from an industrial perspective.

In China, rapid economic growth over the past 40 years has led to an explosive increase in energy demand, leading to a boom in the construction of thermal power generation facilities, and the proportion of aging facilities is rapidly increasing. In particular, as coal has been used for around 70% of energy demands so far, environmental pollution such as acid rain has become serious. In order to prevent such environmental pollution and ensure a stable energy supply, nuclear power generation will be widely introduced. Since nuclear power generation equipment and structural members are subjected to severe operating conditions of high temperatures and long-term loads, higher reliability and safety are required.

In order to extend the life of aging equipment, it is necessary to estimate the remaining life using a more accurate and reliable method at the time of design in order to ensure safety. To this end, it is essential to conduct a detailed analysis of the behavior of creep-fatigue small cracks, which are the substance of damage sustained by high-temperature equipment. Furthermore, if this aging damage caused by creep-fatigue can be annihilated or healed by some method, it will be possible to significantly extend the life of high-temperature equipment, which will be environmentally and economically effective.

Furthermore, at the beginning of the twenty-first century, countries around the world have realized high-efficiency gas turbines and coal gasification power generation that aim to save energy and reduce carbon emissions and each of these devices has been used under temperatures that significantly exceed the operating temperatures of conventional technologies [10, 11]. Regarding the creep-fatigue life under such high temperatures, the most interesting question from an engineering perspective is what temperature and mechanical conditions (load conditions) will result in the shortest life. In order to elucidate this point, rather than conducting long-term creep-fatigue failure tests that require a great deal of effort, it is necessary to conduct analytical tests for crack initiation and growth; i.e., tests that can be analyzed the fracture mechanism, fracture morphology, fracture process, crack mechanics, material structure, environmental effects.

1.2 Research and Problems Regarding Conventional High-Temperature Creep-Fatigue

In research on high-temperature fatigue of metallic materials, Manson [12] and Coffin [13] proposed the so-called Manson–Coffin law in the 1950s, which states that there is a power-law relationship between low-cycle fatigue life and inelastic strain range.

Since then, many experiments have been conducted and many fatigue life estimation equations have been proposed. In particular, various life estimation methods have been proposed to extend these methods to the temperature range where creep deformation occurs at high temperatures. Typical examples are the linear cumulative damage law [14–16], frequency modified fatigue life law [17], and strain range partitioning method [18, 19]. These methods are failure life estimation rules that aim to extrapolate the long-term life and remaining life of high-temperature equipment and structures under actual service conditions based on short-time data obtained in the laboratory. These methods can be used to estimate the strength under creep-fatigue interaction based on the creep stress-rupture time diagram and the fatigue Manson–Coffin law [20]. From a practical point of view, these methods are valid rules of thumb. However, as mentioned above, the deformation and fracture mechanisms of creep vary depending on temperature, stress, strain rate, and stress and strain waveforms, so it is desirable to propose an estimation method based on actual phenomena occurring in the material.

In addition to these experimental estimation methods, fracture mechanics and damage mechanics are based on continuum mechanics to describe fracture behavior over time. Fracture mechanics can be described as crack mechanics in which the unstable fracture initiation point and stable crack propagation rate are evaluated using mechanical quantities representing the stress and strain field or strain rate field near the crack tip. Regarding crack propagation at high temperatures, the propagation of large cracks of several millimeters or more or tens of grains or more can be evaluated based on nonlinear fracture mechanics, typically the J-integral [21]. That is, for creep cracks, the creep J-integral J^* (also called the C^* parameter [22], modified J-integral [23]), and for fatigue cracks, the fatigue J-integral range ΔJ_f (also called the cyclic J-integral [24]), and for creep-fatigue cracks, the creep J-integral range ΔJ_c [25, 26] are the appropriate fracture mechanics parameter [27–31]. A feature of the fracture mechanics evaluation method is that if the test environment is the same as the actual environment, the data from the specimen can be directly applied to the actual machine and is therefore very effective in practical applications, such as cracks at welded joints and stress concentrations. However, creep-fatigue of heat-resistant steel used for long periods of time is characterized by the initiation and growth of multiple spatially distributed small cracks, and material damage caused by the small cracks is the most important issue. Small cracks smaller than a few grain sizes are called microstructurally small crack and cannot be explained by deterministic fracture mechanics assuming homogeneous materials.

On the other hand, damage mechanics was proposed by Kachanov [32] and describes damage such as microscopic defects (cavities and small cracks) in materials using mechanical variables (damage variables, internal variables) from the perspective of continuum mechanics [32–35]. However, the morphology, number, distribution, growth behavior of cracks differ between fatigue, creep, and creep-fatigue. Even under the same creep-fatigue conditions, the mechanisms of deformation and fracture and the dynamics governing them are different between the crack initiation stage at the early life and the large crack propagation stage at the end of life. To improve the accuracy of damage estimation, it is necessary to use appropriate

mechanical variables at each stage. However, since the damage variables in damage mechanics are applied to a global field representing the entire continuum, their relationship to the actual crack morphology and mechanism is unclear and inappropriate, especially under conditions where a small number of cracks dominate the damage or fracture. Recently, attempts have been made to relate damage variables to the amount of damage caused by cavities actually observed (e.g., A-parameter [36, 37], cavity area fraction [36, 37], etc. [38, 39], but it cannot be said that the studies based on geometry and mechanics regarding the amount of damage to be dealt with are sufficient.

To summarize the above, fracture mechanics represents the mechanical state at the tip of a relatively large (long) crack, and the life can be estimated based on the crack. On the other hand, damage mechanics is a method for estimating material damage and life by representing the state of many distributed defects such as small cracks with a single mechanical quantity. Therefore, it is difficult to evaluate remaining life and damage using any of the methods regarding creep-fatigue fracture, where the type and number of defects change depending on load conditions and time even under the same conditions.

1.3 Purpose and Significance of This Book

High-temperature creep-fatigue fracture of austenitic heat-resistant steel is caused by the initiation and growth of intergranular small cracks [2, 8, 40, 41]. However, even among the same intergranular small cracks, in addition to those that form on the material surface as already reported [40–50], there are also those that form inside the material [21, 41]. The initiation and growth of surface small cracks and the resulting fracture under conditions of relatively low temperature, short time, and high stress or high strain rate (referred to as surface cracking type fracture in this book) have been clarified by experiments and analyses [40, 42–50]. However, there has been no systematic study of fractures caused by the initiation, growth, and coalescence of small inner cracks (referred to as inner cracking type fracture), which are considered to appear under conditions of higher temperature, longer time, and lower stress or lower strain rate. In particular, considering that creep-fatigue strength under high-temperature, low-stress, and long-period loading conditions is an issue in actual high-temperature equipment and structures, it is very important to elucidate the initiation and growth behavior of small inner crack.

Many experimental and theoretical studies have been conducted on austenitic stainless steel SUS304 from various viewpoints as a main material for high-temperature equipment in thermal and nuclear power plants [27, 30, 41–57], the mechanism and morphology of high-temperature creep-fatigue fracture are not clearly understood. This may be due to the fact that the fracture mechanism varies depending on the temperature and stress conditions, and that most of the studies have been conducted at relatively low temperatures, high strain rates, or high stress conditions.

Therefore, in this book, the first objective was to experimentally and theoretically clarify the initiation and growth of the creep cavities and small inner cracks in high-temperature creep-fatigue in the austenitic stainless steel SUS304, a typical heat-resistant steel, as the main research material. The second objective was to develop a method for three-dimensionally visualizing and simulating the inner cracking type fracture process and to establish a method for predicting the remaining life of high-temperature components by reproducing the complicated fracture process on a computer. The third objective was to propose a method for annihilating and healing creep-fatigue damage caused by small inner cracks using the results of analyzing inner cracking type fractures and to experimentally prove the method. We believe that the knowledge obtained in this book will make a significant contribution not only to the academic field, but also to the design of various high-temperature equipment used in aircraft, automobiles, chemical and power plants, etc., and to the improvement of technology for predicting the remaining life of the equipment used.

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Chapter 2

Grain Boundary Cavity and Damage Evaluation in Creep-Fatigue



Abstract This chapter presents analytical and experimental results on representative creep damage parameters for grain boundary cavities that initiate under creep-fatigue conditions. First, equations relating creep damage parameters to cavity and grain boundary distributions were derived using a probabilistic model. Then, the size and shape distribution of the grain boundary cavities that initiated under creep-fatigue conditions in SUS304, which is a typical polycrystalline heat-resistant steel, and their change from the initial to the middle of the life of the steel were shown. Based on the observations of cavities, four kinds of creep damage parameters (area fraction of cavities on cross section, fraction of cavities on grain boundary lines, areal cavity density, and A-parameter) were evaluated. For all damage parameters, the damage value increases with increasing life fraction. However, under creep-dominated conditions, where only the grain boundary is preferentially damaged, only the fraction of cavities on grain boundary lines has a clear physical meaning as creep damage.

2.1 Introduction

As mentioned in Chap. 1, heat-resistant materials that are repeatedly subjected to low stress (or low strain rate) loading at high temperatures generally undergo the following three stages of fracture process [1–4].

- (1) A number of creep cavities are initiated on the grain boundary and grow with coalescence.
- (2) Many small cracks, approximately one grain boundary in length, are initiated and propagate with coalescence.
- (3) Preferentially propagated and coalesced small cracks form a large crack, and it propagates following the fracture mechanics law.

In this chapter, the process (1) is explained. From the perspective of ensuring the reliability of high-temperature equipment, it is necessary to evaluate creep damage of used components from the early stage of their service life. Therefore, observation of creep cavities and evaluation of various creep damage parameters closely related to the distribution of cavities have been conducted [5–10]. However, in conventional

studies, there have been very few examples where the comprehensive examination of the shape, size distribution, or temporal changes of individual cavities has been conducted. Additionally, the physical significance of creep damage parameters has not been taken into consideration for investigation. In this chapter, based on the test results conducted under high-temperature fatigue conditions governed by creep using SUS304, which is a representative heat-resistant steel, and the laser microscope observation results of creep cavities on the longitudinal section of interrupted test specimens, the initiation and growth behavior of cavities were examined. Based on microscopic observations [11], various parameters of grip strength impairment were assessed, and the characteristics and physical significance of each parameter over time were discussed.

2.2 Probabilistic Analysis of Creep Damage Parameters

2.2.1 *Creep Damage Parameters*

Cavity Area Fraction [12]

Definition: The percentage of area occupied by cavities on a material surface area or cross-sectional surface area.

Characteristics:

- (i) Only the area of each cavity needs to be measured and can be relatively easily calculated using a microscope and image processing equipment.
- (ii) When cavities are uniformly distributed, the cavity area fraction is equal to the rate of decrease in the weight density of the material. Therefore, the cavity area fraction can also be obtained by measuring the material's density change. However, caution is required in cases where the base material's properties change significantly with each measurement due to factors like oxidation, precipitation, etc.[13].

Areal Cavity Density [4]

Definition: The number of cavities per unit area present on a material surface or cross-sectional surface.

Characteristics: The measurement of the area density of cavities is easy because only the cavities' presence needs to be identified.

Fraction of Cavities on Grain Boundary Lines [5]

Definition: The percentage of the cavity length to the grain boundary length observed on the material surface or cross-sectional surface.

Characteristics: In addition to cavity identification and length measurement, it is necessary to measure the length of the grain boundary, which is not so easy to measure using microscopes, image processing equipment, etc.

A-parameter [14]

Definition: The percentage of grain boundaries where cavities observed among the grain boundaries intersecting the inspection line parallel to the stress axis on the material surface or cross-sectional surface.

Characteristic: The measurement is relatively easy because only the presence or absence of cavities on the grain boundary line is determined.

In addition to the four parameters mentioned above, there are other creep damage parameters such as cavity volume fraction and effective cross-sectional area fraction. In this section, equations relating each parameter to the actual cavity and grain boundary distributions are derived by probabilistic analysis. The A-parameter, cavity volume fraction, and effective cross-sectional area ratio are discussed. First, an analysis concerning the grain boundary planes is conducted, followed by a discussion of the relationship between cavity distribution and creep damage parameters.

2.2.2 Analysis of Grain Boundaries

2.2.2.1 Radius and Angle Distribution of All Grain Boundaries

As shown in Fig. 2.1, all grain boundaries are assumed to be randomly distributed in space and their shape is approximated by a circular disk with no thickness. In actual polycrystalline materials, the grain boundaries are not exactly circular and flat, however, in this analysis, the actual polygonal grain boundary surfaces are simplified and substituted with disks of equal area. The radius of each grain boundary plane is represented by r ($0 < r \leq +\infty$), the position is represented by the coordinate of the center of the grain boundary plane, and the orientation is represented by the surface angle ξ ($-\pi/2 < \xi \leq \pi/2$) and internal angle η ($-\pi/2 < \eta \leq \pi/2$) shown in Fig. 2.2. Here, ξ is the angle between the grain boundary plane and the y -axis in the y - z plane, and η is the angle between the grain boundary plane and the y - z plane. If the intersection of the grain boundary plane and the surface (y - z plane) is called the grain boundary line, then the angle ξ can also be described as the angle between the grain boundary line and the y -axis. Now, let us determine the probability density functions $f_r(r)$, $f_\xi(\xi)$, $f_\eta(\eta)$ for the radius r and angles ξ and η when the grain boundary planes are randomly distributed in space. As defined earlier for the radius r and angles ξ and η , the following are established for grain boundary planes distributed randomly in space.

- (i) The three coordinate components (x_0 , y_0 , z_0) representing the center of the grain boundary plane are independently uniformly distributed.
- (ii) The distributions of r , ξ , and η are independent of each other.
- (iii) The unit normal vector of the grain boundary plane is uniformly distributed on the unit sphere centered at the origin as shown in Fig. 2.3.

Fig. 2.1 Grain boundary plane randomly distributed in space and grain boundary line and inspection line on surface [16]

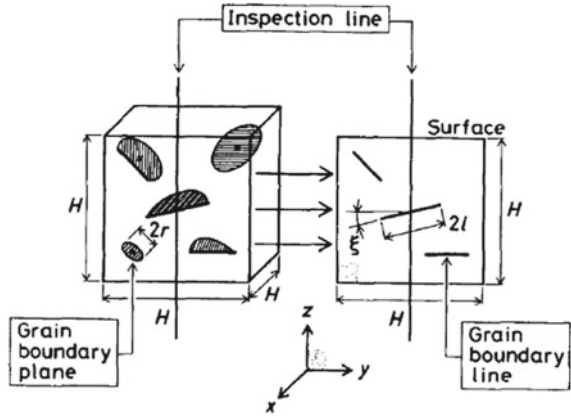
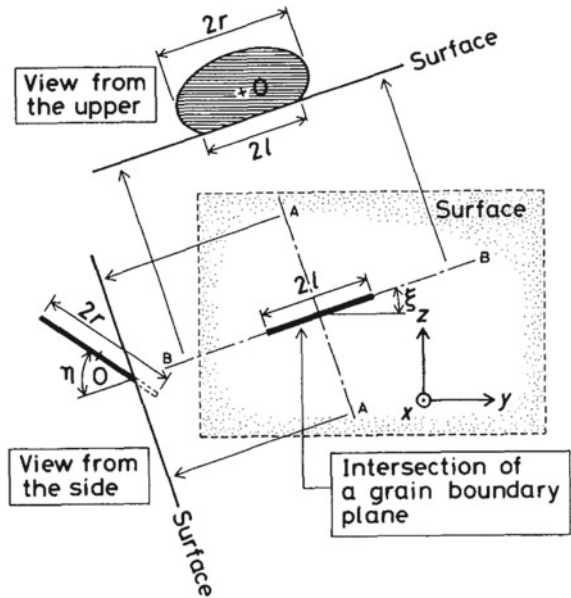
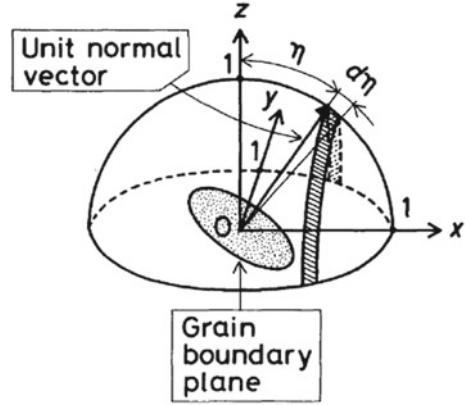


Fig. 2.2 Radius, surface angle, and internal angle of grain boundary plane [16]



From (iii) above, the angle ξ is uniformly distributed in the range of $-\pi/2 < \xi \leq \pi/2$. Therefore, the probability density function $f_{\xi}(\xi)$ is expressed as $1/\pi$. As for the distribution of the internal angle η , the area of the semicircular part of Fig. 2.3 (the shaded area in the figure) is proportional to $\cos \eta d\eta$, and from (iii) above, the probability density function $f_{\eta}(\eta)$ is proportional to $\cos \eta$. Therefore, $f_{\eta}(\eta)$ is expressed as $1/2\cos \eta$ by standardization. The distribution of radius r , which is representative of the size of the grain boundary plane, is arbitrary and independent of the distribution of angles ξ and η , and is not constrained by the random orientation condition of the grain boundary plane. Therefore, when actually determining the

Fig. 2.3 Unit normal vector of grain boundary plane distributed in random orientation [16]



distribution of r for a material, it is necessary to determine the grain boundary line length distribution on the material surface or cross-sectional surface and to determine it using an inverse analytical method [15].

2.2.2.2 Radius and Angle Distributions of Grain Boundary Planes Intersecting a Straight Line

The probability density functions of the radius and angle of the grain boundary planes are calculated only for the grain boundary planes intersecting an arbitrary straight line (line of measurement).

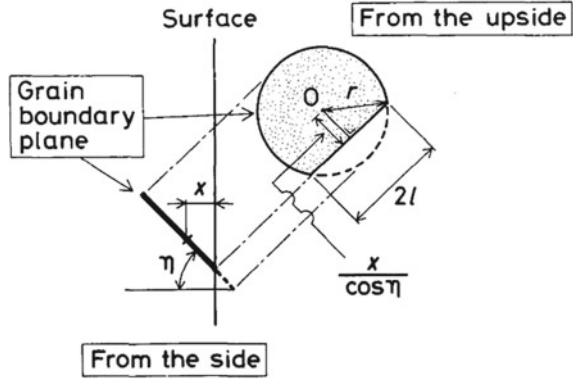
As illustrated in Fig. 2.1, the scenario where a randomly oriented disk-shaped grain boundary plane exists at a random position in a cube of dimension H is considered. In this case, the length of the intersection between the grain boundary plane and the surface, i.e., the total length of the grain boundary line, is $2l$.

(1) Probability of grain boundary plane intersecting a straight line p_l

The case where the radius r , surface angle ξ , and internal angle η are constant is considered. In other words, we focus only on the case where those with the radius, surface angle, and internal angle of r , ξ , and η , respectively, are taken from the population of entire grain boundary planes. The probability p_l that a grain boundary plane of radius r , surface angle ξ , and internal angle η intersects a line parallel to the z -axis is given by the projected area of the grain boundary plane onto the x - y plane is $\pi r^2 \cos \xi \cos \eta$ and is therefore expressed as follows:

$$p_l = \frac{\pi r^2 \cos \xi \cos \eta}{H^2}. \quad (2.1)$$

Fig. 2.4 Relationship between radius of grain boundary plane (r) and grain boundary length (l) and internal angle (η) [16]



(2) Distribution of grain boundary length l

Figure 2.4 shows a grain boundary plane intersecting a surface, viewed from the side (parallel to the grain boundary line) and from the upside (perpendicular to the grain boundary plane). Let x be the distance from the center O of the grain boundary plane to the surface.

- (i) The distance x is distributed in the range of $0 \sim r \cos \eta$.
- (ii) The probability density function $k_x(x)$ of x is proportional to the grain boundary line length l .

As shown in Fig. 2.1, the probability that a straight line on the surface intersects the grain boundary plane, in other words, the probability that the measurement line intersects the grain boundary line, is proportional to the length l of the grain boundary line when r , ξ , and η , are constant ((ii) above). This can be understood from the fact that the number of measurement lines that intersect the grain boundary line depends on the distance x from the center of the grain boundary plane to the surface, which is given in proportion to the length of the grain boundary line, l . Therefore, considering that the distance x , the grain boundary radius r , the grain boundary line length l , and the internal angle η have the relationship in (2.2), the probability density function $k_x(x)$ for x can be expressed as in (2.3) with a standardization.

$$r^2 = l^2 + \left(\frac{x}{\cos \eta} \right)^2$$

$$k_x(x) = A \cdot l \quad (A : \text{Constant}) \quad (2.2)$$

$$k_x(x) = A \cdot \left\{ r^2 - \left(\frac{x}{\cos \eta} \right)^2 \right\}^{1/2}$$

$$= \frac{4l}{\pi r^2 \cos \eta} \cdot \quad (2.3)$$

The probability density function $k_x(x)$ of the grain boundary line length l is obtained through (2.3) and the transformation formula for random variables as follows:

$$\begin{aligned} k_l(l) &= \left| \frac{dx}{dl} \right| \cdot k_x(x) \\ &= \frac{4}{\pi r^2} \cdot \frac{l^2}{(r^2 - l^2)^{1/2}} \end{aligned} \quad (2.4)$$

(3) Simultaneous probability density function $g(l, r, \xi, \eta)$

If the number of grain boundary planes intersecting the measurement line per unit length is n_l and the number of grain boundary planes per unit volume is n_v , the following equation holds for the number of grain boundary planes.

$$\begin{aligned} n_l \cdot H \cdot g(l, r, \xi, \eta) dl dr d\xi d\eta \\ = n_v \cdot H^3 \cdot f_r(r) \cdot f_\xi(\xi) \cdot \zeta_\eta(\eta) \times p_l \cdot k_l(l) dl dr d\xi d\eta. \end{aligned} \quad (2.5)$$

However, $g(l, r, \xi, \eta)$ in the above equation is the simultaneous probability density function with respect to the four random variables l, r, ξ, η . Substituting (2.1) and (2.4) into (2.5), we obtain the following equation for $g(l, r, \xi, \eta)$.

$$g(l, r, \xi, \eta) = \frac{4n_v}{n_l} \cdot \left\{ \frac{l^2 \cdot f_r(r)}{(r^2 - l^2)^{1/2}} \right\} \cdot \{ f_\xi(\xi) \cos \xi \} \cdot \{ f_\eta(\eta) \cos \eta \}. \quad (2.6)$$

Furthermore, integrating the above equation with respect to l, r, ξ, η , we obtain the following equation that expresses the relationship between the number of grain boundary planes n_l and the volume density n_v of the grain boundary planes where the unit length measurement lines intersect.

$$\frac{n_v}{n_l} = \frac{2}{\pi \cdot r_{(m2)}}. \quad (2.7)$$

In the above equation, $r_{(m2)}$ is the mean of the squares of the grain boundary radius r . In general, the i -squared mean of r (i -th moment) is defined by the following equation, which in this study will be expressed by $r_{(mi)}$.

$$r_{(mi)} = \int_0^\infty r^i \cdot f_r(r) dr. \quad (2.8)$$

Substituting (2.7) back into (2.6) and rearranging, we finally obtain the following equation.

$$g(l, r, \xi, \eta) = \frac{4}{\pi^2 \cdot r_{(m2)}} \cdot \left\{ \frac{l^2 \cdot f_r(r)}{(r^2 - l^2)^{1/2}} \right\} \cdot \cos \xi \cdot \cos^2 \eta \quad (2.9)$$

$$g_{l,r}(l, r) = \frac{4}{\pi \cdot r_{(m2)}} \cdot \left\{ \frac{l^2 \cdot f_r(r)}{(r^2 - l^2)^{1/2}} \right\} \quad (2.10)$$

$$g_\xi(\xi) = \frac{1}{2} \cdot \cos \xi \quad (2.11)$$

$$g_\eta(\eta) = \frac{2}{\pi} \cdot \cos^2 \eta. \quad (2.12)$$

As indicated in the above equation, the radius r and the grain boundary line length l of the grain boundary plane are not independent and have the relationship in (2.10). On the other hand, they are independent of each other from the surface angle ξ and the internal angle η .

2.2.3 Relationship Between Cavity Distribution and Creep Damage Parameters

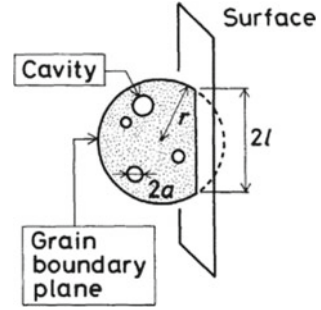
Cavity volume fraction, effective cross-sectional area fraction, and A-parameter among the creep damage parameters are discussed in relation to the actual distribution of cavities.

Cavity Volume Fraction V_c

This section discusses the relationship between the cavity volume fraction V_c (the total volume of cavities per unit volume) and the actual cavity distribution. First, all cavities are assumed to be spherical in shape. This assumption is a good approximation in many cases, such as when cavity dimensions are small and surface diffusion processes dominate, or when cavities originate at the interface between spherical inclusions and the base metal, where the adhesion to the base metal is not good.

First, cavities on a grain boundary plane of radius r , surface angle ξ , and internal angle η are considered. The case where many spherical cavities of radius a [probability density function $f_a(a)$] are distributed on a grain boundary plane is considered as shown in Fig. 2.5. The following is an example of a spherical cavity with radius a [probability density function $f_a(a)$]. If the area density of cavities on the grain boundary plane of interest (the number of cavities on the grain boundary plane in a unit area) is n_{cs} , the total volume V_c of cavities on the grain boundary plane can be expressed as follows:

Fig. 2.5 Several creep cavities and grain boundary lines on a single grain boundary plane [16]



$$\begin{aligned}
 v_c(r, \xi, \eta) &= n_{cs} \cdot \int_0^\infty \frac{4}{3} \pi \cdot a^3 \cdot f_a(a) da \cdot \pi r^2 \\
 &= \frac{4}{3} \pi^2 \cdot n_{cs} \cdot a_{(m3)} \cdot r^2.
 \end{aligned} \tag{2.13}$$

It is assumed that n_{cs} and $a_{(m3)}$ in the above equation are independent of the radius of the grain boundary plane and are expressed as functions of the angles ξ and η with respect to the stress axis of the surface. That is, $n_{cs} = n_{cs}(\xi, \eta)$ and $a_{(m3)} = a_{(m3)}(\xi, \eta)$. This assumption implies that the resistance to cavity initiation and growth at grain boundary planes is constant regardless of the position and orientation of the grain boundary planes when the effect of grain boundary structure, i.e., loading direction, is removed. Therefore, this assumption is believed to hold true for most isotropic polycrystalline materials.

Integrating the above equation for all grain boundaries [radial distribution is $f_r(r)$, surface angular distribution is $f_\xi(\xi)$, and internal angular distribution is $f_\eta(\eta)$], we obtain the following equation.

$$\begin{aligned}
 V_c &= n_{fv} \cdot \int_{-\infty}^{\infty} \int_0^{\pi/2} \int_{-\pi/2}^{\pi/2} \int_{-\pi/2}^{\pi/2} v_c \times f_r(r) f_\xi(\xi) f_\eta(\eta) d\eta d\xi dr \\
 &= \frac{2}{3} \pi \cdot n_{fv} \cdot r_{(m2)} \times \int_{-\pi/2}^{\pi/2} \int_{-\pi/2}^{\pi/2} n_{cs} \cdot a_{(m3)} \cdot \cos \eta d\eta d\xi
 \end{aligned} \tag{2.14}$$

n_{fv} in the above equation is the volume density of the grain boundary plane (the number of grain boundary planes per unit volume). This indicates that the cavity volume fraction V_c is a quantity proportional to the product of the area density of cavities on the grain boundary and the cubic of the cavity radius, when the effect of the grain boundary plane distribution is removed.

Effective Cross-sectional Area Ratio

The definition of effective cross-sectional area in damage mechanics is not entirely clear, and it is treated qualitatively or conceptually as the actual cross-sectional area that supports the load. Therefore, the effective cross-sectional area ratio (ratio of total cross-sectional area to net cross-sectional area) is interpreted in two ways, and the relationship between each and cavity distribution is analyzed.

- (1) Cavity area ratio for all grain boundary planes (effective cross-sectional area ratio, part 1)

We defined the effective cross-sectional area fraction S_{clf} as the ratio of the non-cavity area to the grain boundary area for all grain boundary planes. As before, we first consider the case where the radius r , surface angle ξ , and internal angle η are constant. Let s_c be the area occupied by the cavity on the grain boundary plane, and s_c is given by

$$\begin{aligned} s_c &= n_{cs} \cdot \int_0^{\infty} \pi a^2 \cdot f_a(a) da \cdot \pi r^2 \\ &= \pi^2 \cdot n_{cs} \cdot a_{(m2)} \cdot r^2. \end{aligned} \quad (2.15)$$

Integrating the above equation over the entire grain boundary plane in the unit volume gives the total cavity area, s_c , over all the grain boundary planes in the unit volume.

$$\begin{aligned} S_c &= n_{fv} \cdot \int_0^{\infty} \int_{-\pi/2}^{\pi/2} \int_{-\pi/2}^{\pi/2} s_c \times f_r(r) f_{\xi}(\xi) f_{\eta}(\eta) d\eta d\xi dr \\ &= \frac{\pi}{2} \cdot n_{fv} \cdot r_{(m2)} \times \int_{-\pi/2}^{\pi/2} \int_{-\pi/2}^{\pi/2} n_{cs} \cdot a_{(m2)} \cdot \cos \eta d\eta d\xi. \end{aligned} \quad (2.16)$$

On the other hand, the total area of the grain boundary plane in a unit volume, S_f , is expressed as in (2.17), and thus, the effective cross-sectional area ratio S_{clf} is expressed as in (2.18).

$$S_f = \pi \cdot n_{fv} \cdot r_{(m2)} \quad (2.17)$$

$$\begin{aligned} S_{c/f} &= 1 - \frac{S_c}{S_f} \\ &= 1 - \frac{1}{2} \cdot \int_{-\pi/2}^{\pi/2} \int_{-\pi/2}^{\pi/2} n_{cs} \cdot a_{(m2)} \cdot \cos \eta d\eta d\xi. \end{aligned} \quad (2.18)$$

(2) Cavity area ratio in the case of only vertical grain boundary planes (effective cross-sectional area ratio, part 2)

The population of the grain boundary plane of interest is the grain boundary plane in the direction near perpendicular to the stress axis (z -axis). In the case of uniaxial creep or in the case of multiaxial creep where one of the principal stresses is dominant, creep cavities and small cracks caused by the growth and linkage of these cavities are concentrated and grow at the grain boundary plane perpendicular to the stress axis. Therefore, grain boundaries parallel to the stress axis are less damaged and do not need to be considered as effective grain boundaries because they do not provide a fracture path (crack propagation path). The effective cross-sectional area fraction in (1) better represents the damage in materials subjected to strong multiaxial loading, while the effective cross-sectional area fraction considered in (2) better represents creep damage in materials subjected to strong uniaxial loading. However, in cases where the grain boundary planes is not the path of fracture, i.e., intragranular fracture types, the cavity volume fraction is a more appropriate quantity to represent the damage.

Numerous methods exist for identifying grain boundaries that are perpendicular to the stress axis. Here we define the perpendicular grain boundary planes as those that intersect the measurement line (a straight line parallel to the stress axis) and define them as the population. In this case, the cavity occupied area $\tilde{s}_c$ at a certain vertical grain boundary plane (assuming constant radius r , surface angle ξ , and internal angle η) is expressed as follows, as in (2.15).

$$\tilde{s}_c = \pi^2 \cdot n_{cs} \cdot a_{(m2)} \cdot r^2.$$

Also, the total cavity area on the grain boundary plane intersecting the unit length measurement line, $\tilde{S}_c$ is expressed as follows:

$$\begin{aligned} \tilde{S}_c &= n_{fl} \cdot \int_0^\infty \int_{-\pi/2}^{\pi/2} \int_{-\pi/2}^{\pi/2} \tilde{s}_c \times g_r(r) g_\xi(\xi) g_\eta(\eta) d\eta d\xi dr \\ &= \pi \cdot n_{fl} \cdot \frac{r_{(m4)}}{r_{(m2)}} \times \int_{-\pi/2}^{\pi/2} \int_{-\pi/2}^{\pi/2} n_{cs} \cdot a_{(m2)} \cdot \cos \xi \cdot \cos^2 \eta d\eta d\xi. \end{aligned} \quad (2.19)$$

Therefore, the total area $\tilde{S}_f$ of the grain boundary plane intersecting the unit length line and the effective cross-sectional area ratio $\tilde{S}_{c/f}$ at the vertical grain boundary plane are obtained as follows.

$$\tilde{S}_f = \pi \cdot n_{fl} \cdot \frac{r_{(m4)}}{r_{(m2)}} \quad (2.20)$$

$$\begin{aligned}
\tilde{S}_{c/f} &= 1 - \tilde{S}_c / \tilde{S}_f \\
&= 1 - \int_{-\pi/2}^{\pi/2} \int_{-\pi/2}^{\pi/2} n_{cs} \cdot a_{(m2)} \cdot \cos \xi \cdot \cos^2 \eta d\eta d\xi.
\end{aligned} \tag{2.21}$$

As evident from (2.18) and (2.21) above, both effective cross-sectional area fractions are essentially the quantity 1 minus the product of the cavity area density on the grain boundary plane and the square of the cavity radius. However, the value of $\tilde{S}_{c/f}$ expressed in (2.21) is generally smaller than the value of $S_{c/f}$ expressed in (2.18) because of the stronger dependence of angles ξ and η . For uniaxial creep tests and loading conditions in actual machines where large maximum principal stresses are applied, $\tilde{S}_{c/f}$ in (2.21) is considered to more accurately represent creep-cavity damage than $S_{c/f}$ in (2.18).

A-parameter

The A-parameter is defined as the percentage of grain boundary lines with one or more cavities that intersect the measurement line parallel to the stress axis. In this section, we discuss the relationship between the A-parameter A and cavity distribution.

The cavity [radius distribution $f_a(a)$] on a grain boundary plane with radius r , surface angle ξ , and internal angle η as shown in Fig. 2.5 is considered. The relationship between the cavity area density on the grain boundary n_{cs} and the number of cavities per unit length n_{cl} (density of cavities at grain boundary line) can be expressed by the following equation [15].

$$n_{cl} = 2n_{cs} \cdot a_{(m1)}. \tag{2.22}$$

Thus, the number of cavities N_{cl} appearing at the grain boundary line of length $2l$ is expressed by the following equation.

$$\begin{aligned}
N_{cl} &= 2l \cdot n_{cl} \\
&= 4 \cdot l \cdot n_{cs} \cdot a_{(m1)}.
\end{aligned} \tag{2.23}$$

Here, the A-parameter A is obtained as the average value of N_{cl} when the vertical grain boundary line is the population. However, if multiple cavities frequently appear on the same grain boundary line, it is necessary to correct the value of N_{cl} above one to 1 to obtain the average value.

$$\begin{aligned}
A &= N_{cl(m1)} \\
&= 4 \cdot \int_0^\infty \int_{-\pi/2}^{\pi/2} \int_{-\pi/2}^{\pi/2} l \cdot n_{cs} \cdot a_{(m1)} \times g_l(l) g_\xi(\xi) g_\eta(\eta) d\eta d\xi dl
\end{aligned}$$

$$= \frac{32}{3\pi^2} \cdot \frac{r_{(m3)}}{r_{(m2)}} \times \int_{-\pi/2}^{\pi/2} \int_{-\pi/2}^{\pi/2} n_{cs} \cdot a_{(m1)} \cdot \cos \xi \cos^2 \eta d\eta d\xi. \quad (2.24)$$

While the A-parameter seems to consider only the number of cavities because of the measurement methodology, as evident from the above equation, the value of A is expressed in the form of the product of the cavity area density on the grain boundary plane and the cavity radius, which includes both the number and size of cavities. It can also be seen that the value of A depends on the radial distribution of the grain boundary plane of the material, i.e., the grain size distribution. Therefore, it is necessary to be careful when comparing A not only between different materials, but also between the same material with different grain sizes and distributions due to heat treatment, etc.

The original definition of the A-parameter is the ratio of the area (= number ratio) of cavity grain boundary plane, i.e., the percentage of grain boundary planes where on or more cavities occur among the grain boundary planes perpendicular to the stress axis (the group of grain boundary planes that intersect a line parallel to the stress axis, including internal grain boundary planes) [15]. This is based on the idea that if even one cavity occurs at a grain boundary plane, the grain boundary plane is considerably damaged, and the damage is the same as if a grain boundary-sized microcrack had occurred there. Deriving the value of the A-parameter $\tilde{A}$ according to this original definition can be derived using a similar approach as the equation mentioned above.

$$\tilde{A} = \frac{r_{(m6)}}{r_{(m4)}} \cdot \int_{-\pi/2}^{\pi/2} \int_{-\pi/2}^{\pi/2} n_{cs} \cos \xi \cos^2 \eta d\eta d\xi. \quad (2.25)$$

This is distinct from the value in (2.24) and represents only the area density of cavities or number of cavities. It is very difficult to actually measure the value of $\tilde{A}$, and in practice, A is often measured as an indicator of cavity damage. However, it is important to recognize that the two values are fundamentally different.

2.3 Creep-Fatigue Tests

2.3.1 Material, Specimen, and Test Method

The test material was an austenitic stainless steel SUS304, and its chemical composition is listed in Table 2.1. The shape of specimen was round bar as shown in Fig. 2.6. The specimen has a parallel part of 10 mm in diameter and 20 mm in length at the center. The surface of the parallel part was polished with No. 3000 emery paper in the direction of the stress axis before starting the creep-fatigue test to remove

Table 2.1 Chemical composition

(wt.%)									
C	Si	Mn	P	S	Cu	Ni	Cr	Mo	Fe
0.06	0.31	1.07	0.31	0.018	0.44	8.68	18.37	0.25	bal

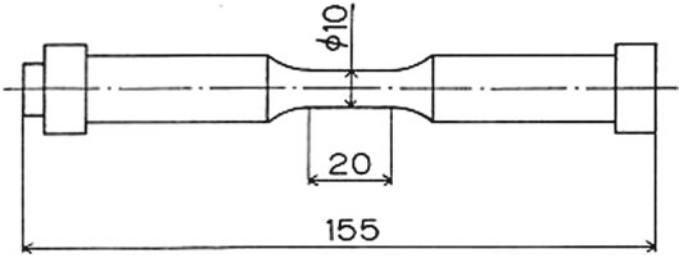
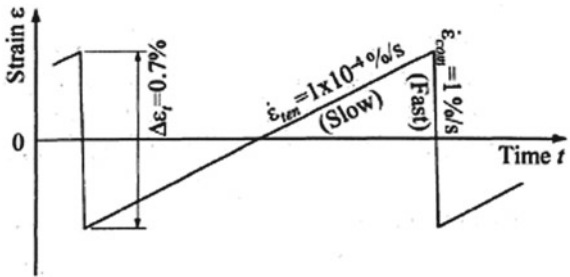


Fig. 2.6 Round bar specimen for creep-fatigue test [11]

Fig. 2.7 Strain waveform [11]



any machining defects. Cavities were observed and measured inside the parallel part. Creep-fatigue tests were conducted using an electric-hydraulic servo fatigue testing machine equipped with a high-frequency induction heater. The experiment was conducted at a temperature of 1073 K (800 °C) in air under strain control. The strain waveform used was the asymmetric triangular slow-tensile fast-compressive (slow-fast) wave shown in Fig. 2.7. The test conditions were set at 1073 K, which is considerably higher than the working temperature of Type 304, in order to select conditions [4, 17], where the specimen generates a large number of internal cavities in combination with the strain waveform. The failure life (the number of fatigue cycles when the maximum stress value decreases to 3/4 of the steady state) under this test condition is 133 cycles. In this chapter, four interrupted specimens with 13, 22, 33, and 66 cycles, corresponding to stage (1) described in the introduction, were prepared for cavity observation. Each specimen is referred to as a $1/10N_f$, $1/6N_f$, $1/4N_f$, or $1/2N_f$ specimen. The internal microcrack initiation and propagation behavior in the mid- to late-life stage, which correspond to stages (2) and (3), will be discussed in later chapters.

2.3.2 Method of Measuring Cavity Shape and Dimensions

After the creep-fatigue test was interrupted, the parallel part of the specimen was split longitudinally near the center and the cross section was polished using No. 2000 emery paper. The longitudinal section was then buffed with diamond paste of approximately 1 μm in grain diameter, and the adhesion and removal of an acetyl cellulose replica film were repeated five times to remove small adhesions such as dust on the observed cross section. Therefore, precipitates with poor adhesion to the base metal were removed by this process, and the holes where precipitates were removed were regarded as cavities in the later observation. In this study, such precipitates with low interfacial strength are considered to cause the same degree of damage to the material as cavities and are treated without distinguishing them from cavities.

A scanning laser microscope was used for observation and measurement for cavity dimensions. This microscope is equipped with a confocal optical system employing a He-Ne laser as a light source and an image analysis system. Therefore, two-dimensional (planar) measurements including the vertical and horizontal dimensions of the object can be easily performed on a monitor, and the thickness and surface profile of the object. Table 2.2 shows the region on the longitudinal cross section of the specimen where cavity measurements were conducted. The area differs for each specimen because a larger measurement area was used when the number of cavities distributed on the longitudinal cross section was limited to obtain statistically more significant results.

Since creep cavities generally occur on grain boundary lines (actually grain boundary planes). Consequently, in this study, measurements of grain boundary line length and angles on the longitudinal cross section were carried out concurrently. It's worth noting that the measurement of grain boundary lines was performed by observing numerous precipitates aligned along them using a laser microscope, and it was separately confirmed that the grain boundary lines measured from these precipitates were the same as those observed after normal etching treatment (e.g., immersion in a solution of hydrofluoric acid, nitric acid, and glycerol). Since all grain boundary lines in this material have an almost linear shape, the total grain boundary line (dashed line in Fig. 2.8) was approximated by a straight line connecting triple points of grain boundary (solid line in Fig. 2.8), and the length (half-length) l_g and the angle θ_g to the vertical direction of the stress axis were evaluated by the approximated grain boundary lines. Specifically, the component $2l_y$ along the stress axis and the component $2l_x$ perpendicular to the stress axis of the total length $2l_g$ of the grain boundary

Table 2.2 Area for cavity observation on the cross section

Life fraction, N/N_f	Observation area, S (mm^2)
1/10	4.43
1/6	3.11
1/4	1.83
1/2	1.19

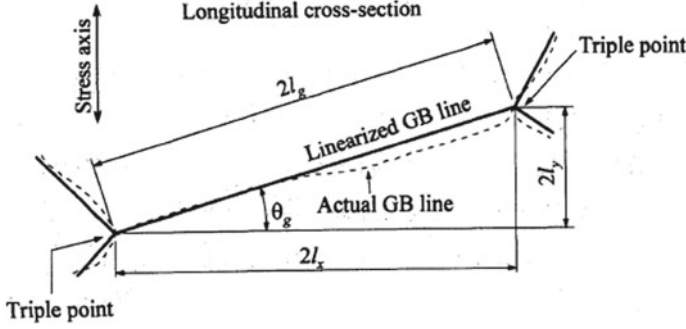


Fig. 2.8 Grain boundary line approximated by a straight line between two triple points [11]

line were measured, and the grain boundary length l_g and angle θ_g were calculated from them by the following equations.

$$l_g = \sqrt{l_x^2 + l_y^2} \quad (2.26)$$

$$\theta_g = \tan^{-1}\left(\frac{l_y}{l_x}\right) \quad (0 \leq \theta_g \leq \frac{\pi}{2}). \quad (2.27)$$

Figure 2.9 illustrates a schematic diagram of cavities on the grain boundary line. In this study, each cavity's shape and dimensions were evaluated by approximating them with ellipses, using the grain boundary direction and the direction perpendicular to it as axes. For example, as shown in Fig. 2.9a, if the total lengths of the cavities on the grain boundary line at the angle θ_g are $2y_0$ and $2x_0$ in the direction of the stress axis and perpendicular to it, respectively, the length (half-length) a_c and the height (half-height) h_c of the cavity along the grain boundary line are given by The length of the cavity along the grain boundary line (half-length) a_c and the height perpendicular to it (half-height) h_c can be calculated from the following equations:

$$a_c = \sqrt{\frac{x_0^2 \cos^2 \theta_g - y_0^2 \sin^2 \theta_g}{\cos 2\theta_g}} \quad \left(0 \leq \theta_g \leq \frac{\pi}{8}, \frac{3\pi}{8} \leq \theta_g \leq \frac{\pi}{2}\right) \quad (2.28)$$

$$h_c = \sqrt{\frac{y_0^2 \cos^2 \theta_g - x_0^2 \sin^2 \theta_g}{\cos 2\theta_g}} \quad \left(0 \leq \theta_g \leq \frac{\pi}{8}, \frac{3\pi}{8} \leq \theta_g \leq \frac{\pi}{2}\right). \quad (2.29)$$

The cavity length a_c and height h_c were obtained from (2.28) and (2.29) by measuring the angle θ_g of the grain boundary line where each cavity exists and the total width $2y_0$ and $2x_0$ of the cavity in the vertical and horizontal directions using a laser microscope. However, for cavities located on the grain boundary line where the grain boundary angle θ_g is near 45° , the application of (2.28) and (2.29) results in a large calculation error, so instead of widths $2y_0$ and $2x_0$, the vertical dimension $2y_{45}$

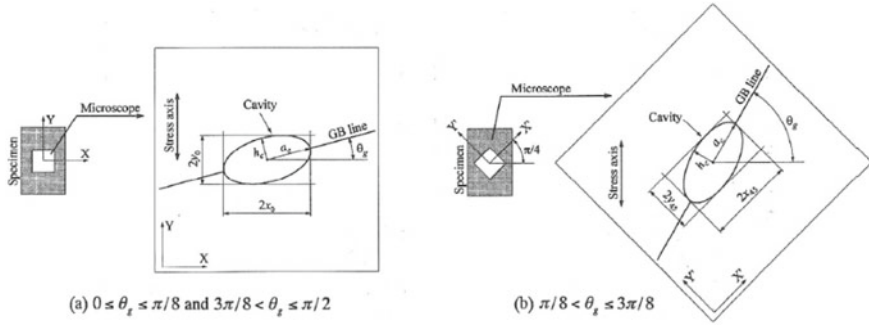


Fig. 2.9 Evaluation of the size of an elliptical cavity lying on a grain boundary line the angle of which is θ_g [11]

and the horizontal dimension $2x_{45}$ were measured when the specimen was viewed with a 45° rotation, and the cavity length and dimensions were evaluated using the following equations instead of (2.28) and (2.29).

$$a_c = \sqrt{\frac{x_{45}^2 \cos^2(\theta_g - \frac{\pi}{4}) - y_{45}^2 \sin^2(\theta_g - \frac{\pi}{4})}{\cos 2(\theta_g - \frac{\pi}{4})}} \quad \left(\frac{\pi}{8} < \theta_g \leq \frac{3\pi}{8} \right) \quad (2.30)$$

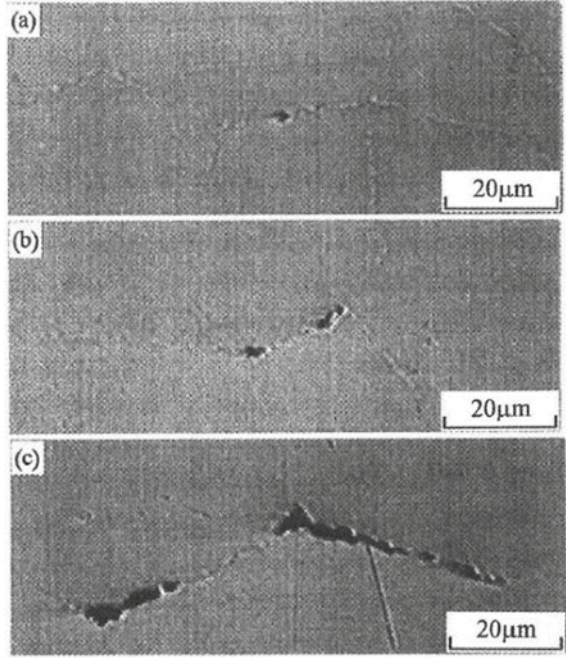
$$h_c = \sqrt{\frac{y_{45}^2 \cos^2(\theta_g - \frac{\pi}{4}) - x_{45}^2 \sin^2(\theta_g - \frac{\pi}{4})}{\cos 2(\theta_g - \frac{\pi}{4})}} \quad \left(\frac{\pi}{8} < \theta_g \leq \frac{3\pi}{8} \right) \quad (2.31)$$

2.4 Distribution of Cavities Generated by Creep-Fatigue

2.4.1 Distribution of Cavity Dimensions and Shapes

Figure 2.10 shows an example of a cavity observation using a laser microscope. The black areas in the photograph are cavities on the grain boundary. The distribution of cavity length and height for each interrupted cycle is shown in Fig. 2.11. Although the length and height of the cavities gradually increase from $1/10$ to $1/4 N_f$, most of the cavities have dimensions of approximately $2\text{--}3 \mu\text{m}$ or less and the length-to-height ratio of the cavities varies around 1. In other words, the cavities are slightly distorted and spherical. On the other hand, at $1/2 N_f$ in the middle of the lifetime, cavities whose length is larger than their height begin to appear. The cavities are spherical in shape when they initiate, but after the initiation, they preferentially grow and coalesce along the grain boundary line, and their shape shifts to flattened cavities that are long in the grain boundary line direction.

Fig. 2.10 Cavities observed by a scanning laser microscope at $N/N_f = 1/2$ [11]



2.4.2 Characteristics of Grain Boundaries Where Cavities Exist

Figure 2.12 shows the length distribution of grain boundary lines in this material. From this figure, it can be observed that the lengths of crystal grain boundaries vary widely, ranging from below $10\text{ }\mu\text{m}$ to several tens of μm . It is worth noted that this data was obtained from measurements conducted on a virgin material subjected to 24 h of heating in air at 1073 K (800 °C), and the results from this study and the distribution of grain boundaries measured on $1/2N_f$ test specimens are confirmed to be almost identical. The angular distribution of the grain boundary lines is shown in Fig. 2.13. As can be inferred from the figure, the angles of the grain boundary lines are almost uniformly distributed from 0 to $\pi/2$. This indicates that there is no bias in the orientation of the grain boundaries and that the specimen is an isotropic material.

The length distribution of the grain boundary line where cavities were observed is shown in Fig. 2.14. In this study, the grain boundary line where cavities are observed on the longitudinal section is referred to as the “damaged grain boundary line”. Comparing the length distribution of the damaged grain boundary lines with that of the dashed line in the same figure, the length distribution of the damaged grain boundary lines is slightly on the longer side. In other words, cavities are often observed on long grain boundary lines. However, considering this in a three-dimensional context, as illustrated schematically in Fig. 2.15, the probability of

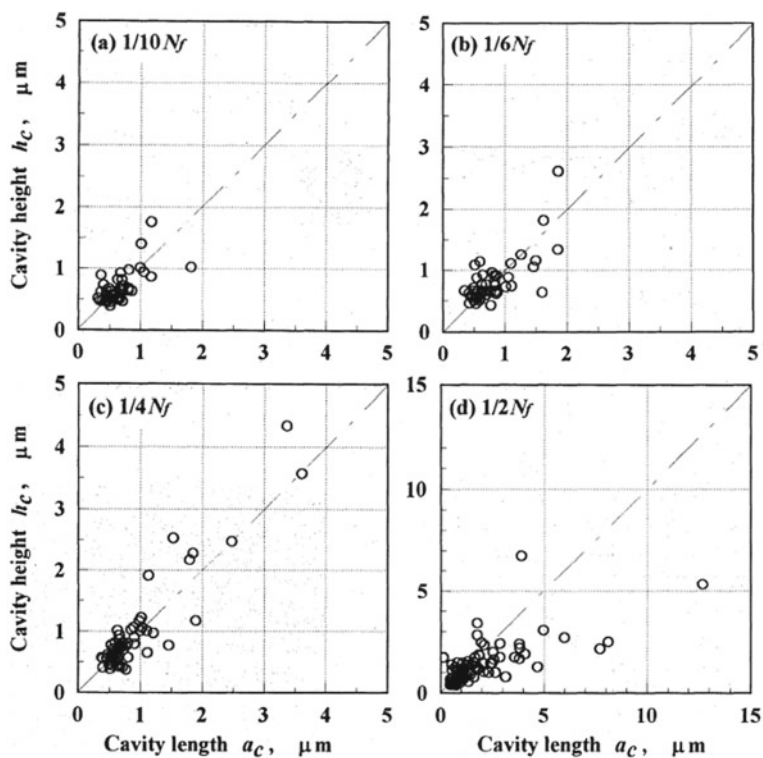


Fig. 2.11 Distribution of the length and height of cavities [11]

Fig. 2.12 Distribution of the length of grain boundary lines [11]

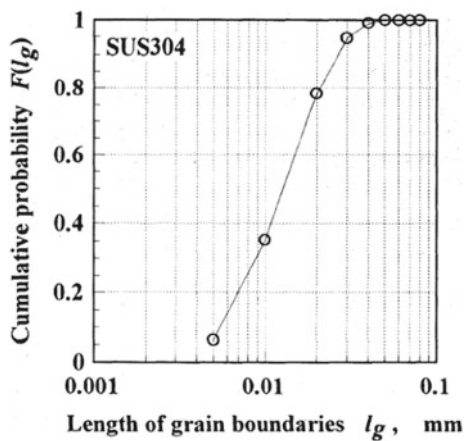
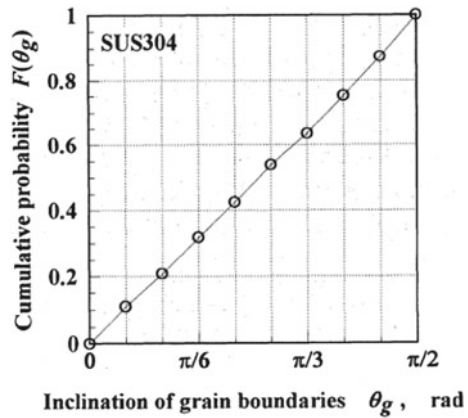


Fig. 2.13 Distribution of the angle of grain boundary lines [11]



finding a grain boundary cavity on a grain boundary line in the longitudinal section is proportional to the length of the grain boundary line. Consequently, if grain boundary damage is determined by the presence or absence of cavities on the grain boundary line, the length of the damaged grain boundary line will consistently lean toward being longer than that of the total grain boundary line. The average linear density of cavities on the grain boundary line (average number of cavities on the grain boundary line per unit length) was calculated for the $1/2N_f$ specimen with the highest average number of cavities on a grain boundary line among the four interrupted specimens, and the results were classified by grain boundary line length. The results are classified by grain boundary length and are shown in Table 2.3. The results are categorized by grain boundary length and are shown in Table 2.3. The results are almost constant, ranging from 3 to 4 (mm^{-1}), regardless of the size of the grain boundary and the length of the exposed grain boundary line (the location of the grain boundary cut). In other words, the number of cavities does not depend on the size of the grain boundary or its position on the grain boundary, and the reason why cavities appear to occur preferentially at longitudinal sections at long grain boundary lines is that the probability of cavities being exposed on the longitudinal section differs between long grain boundary lines and short grain boundary lines. It can be concluded that this is because the probability of cavities being exposed on the longitudinal section differs between long and short grain boundary lines which indicates that when assessing creep damage in materials due to the presence of cavities, one should always pay attention to the relationship between the damage observed in two dimensions and the actual damage.

The angular distribution of the damaged grain boundary line relative to the vertical direction of the stress axis is shown in Fig. 2.16. Comparing the angular distribution of the damaged grain boundary line with the angular distribution of the total grain boundary line shown by the dashed line in this figure, the angular distribution of the damaged grain boundary line is uniform as well as that of the total grain boundary line, because the two are identical up to $1/4N_f$ specimens. However, in the mid-life $1/2 N_f$ specimen, cavities preferentially exist in the direction perpendicular to the

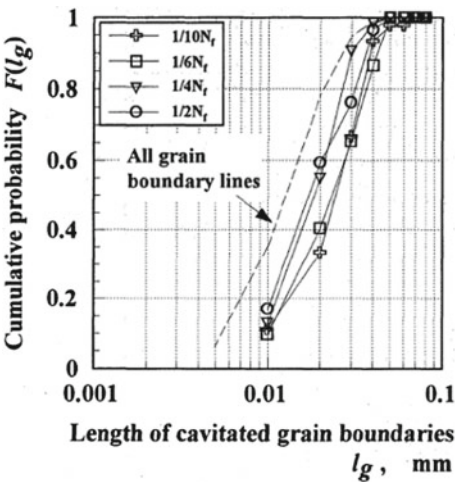


Fig. 2.14 Distribution of the length of cavitated (or damaged) grain boundary lines [11]

Fig. 2.15 A schematic figure showing the relationship between the length of a grain boundary line and the number of cavities appeared on the line [11]

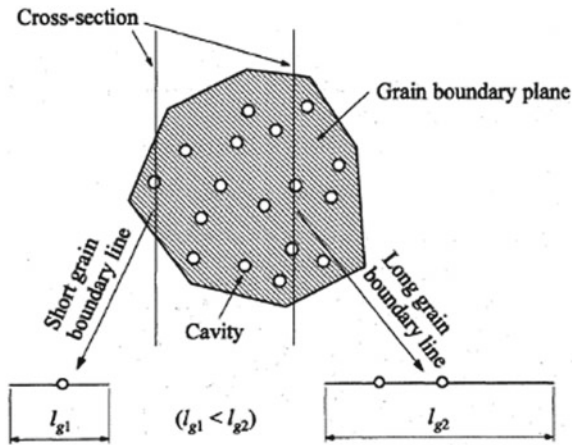


Table 2.3 Density of cavities on grain boundary lines at $N/N_f = 1/2$

Length of grain boundary lines, l_g (mm)	$l_g < 0.01$	$0.01 \leq l_g < 0.02$	$l_g \geq 0.02$
Density of cavities on GB lines n_{cl} (mm^{-1})	3.50	3.64	3.70

stress axis. The relationship between the length of the cavities and the angle of the grain boundary line where the cavities existed at each interruption repetition rate is shown in Fig. 2.17.

Figure 2.17 shows the relationship between the length of the cavity and the angle of the grain boundary line where the cavity existed at each number of interruption cycles.

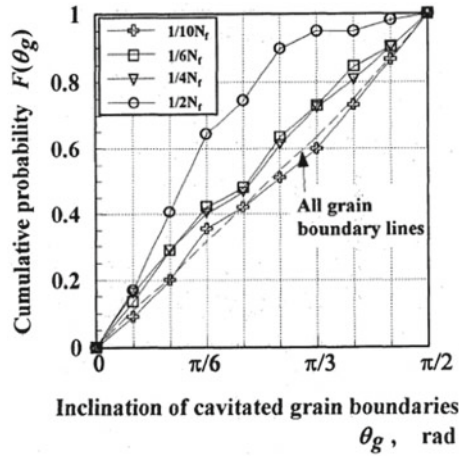


Fig. 2.16 Distribution of the angle of cavitated (or damaged) grain boundary lines [11]

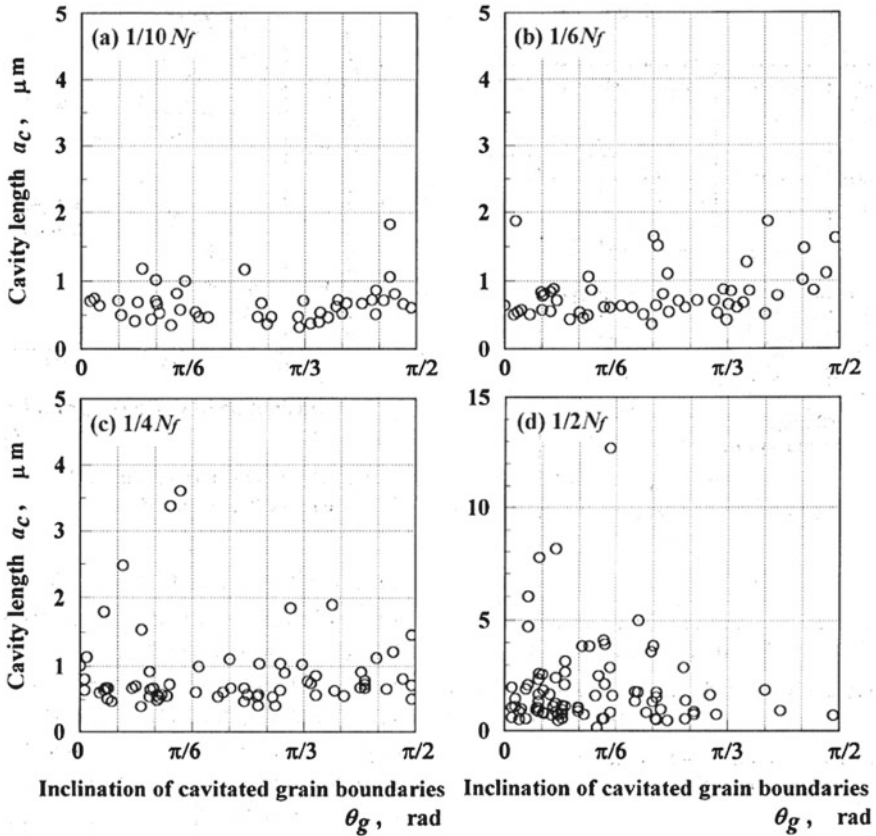


Fig. 2.17 Relationship between the length of cavities and the angle of grain boundary lines [11]

As shown in this figure, relatively small cavities of about 2–3 μm or less exist on random grain boundary lines regardless of the orientation of the grain boundary lines, while larger cavities tend to exist biasedly on grain boundary lines perpendicular to the stress axis. In other words, the occurrence of cavities is independent of the grain boundary line (plane) orientation, but the growth and coalescence of cavities strongly depend on the orientation of the grain boundary line (plane) in which the cavities exist relative to the stress axis.

2.4.3 Evaluation of Creep Damage Parameters

Various damage parameters have been proposed to quantitatively evaluate damage caused by creep cavity [2, 5, 10], in this study, four types of parameters shown in Fig. 2.18 were evaluated based on the results of measurement of the cavity dimensions and shapes.

The cavity area ratio P is the ratio of the cavity area to the observed area and is expressed by the following equation, where the area of the observed area is S , the cavity length is a_c , and the cavity height is h_c .

$$P = \frac{\sum \pi a_c h_c}{S}. \quad (2.32)$$

The change in cavity area fraction P is shown in Fig. 2.19a. Since this parameter represents the volume fraction of cavities inside the material, it increases in proportion to the product of the number of cavities (volume density) and the cubic of the cavity dimension (e.g., radius), which increases with time [16, 18] when the cavity shape does not change significantly. Thus, its rate of increase is relatively large and sensitive to the progression of damage caused by cavities. However, this parameter does not take into account the existence of grain boundaries. Thus, this parameter may be effective in cases where fracture is caused by interconnected voids that exist independently of grain boundaries, such as intragranular voids in ductile fracture. However, when the fracture is concentrated only at the grain boundary, as in the case in this condition, the meaning of this parameter is ambiguous and its application can be problematic.

The area density of cavities, M , is the number of cavities per unit area in the observation area. When the area of the observation area is S and the number of cavities observed in it is m , the area density is expressed by the following equation.

$$M = \frac{m}{S}. \quad (2.33)$$

Figure 2.19b shows the change of cavity area density M over time. The area density provides information about the number of cavities observed on the longitudinal section, which is proportional to the product of the number of cavities inside the

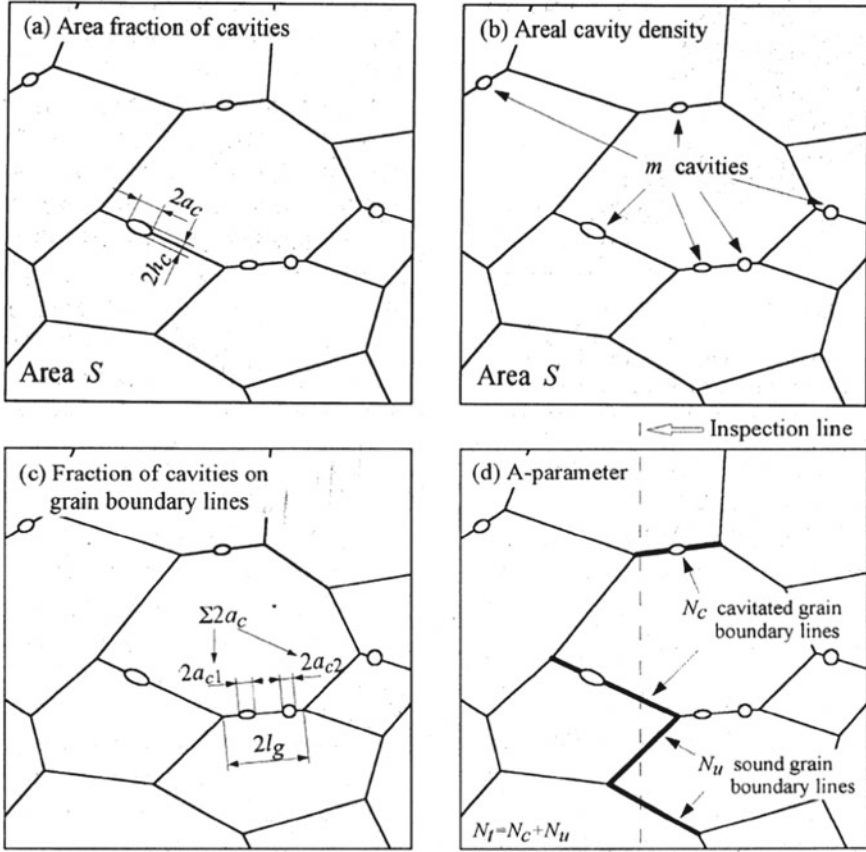


Fig. 2.18 Four kinds of creep damage parameters [11]

material (volume density) and the cavity dimensions (e.g., radius) [16, 18]. Therefore, the rate of increase in the cavity area fraction, which is proportional to the product of the number and the cube of the dimensions of the cavities, is not very high. However, under the present test conditions, the parameter M is increasing almost proportionally to the increase in the fatigue life ratio. Nevertheless, the cavity area density has limited physical meaning, and it remains questionable whether it accurately represents the creep damage induced by the occurrence and growth of cavities along grain boundary planes

The fraction of cavities on grain boundary lines, L , represents the proportion of length occupied by cavities on each grain boundary line and can be expressed using the length of each grain boundary line, l_g , and the total length of cavities on that grain boundary line, Σa_c , as shown in the following equation.

$$L = \left[\frac{\Sigma a_c}{l_g} \right]_m. \quad (2.34)$$

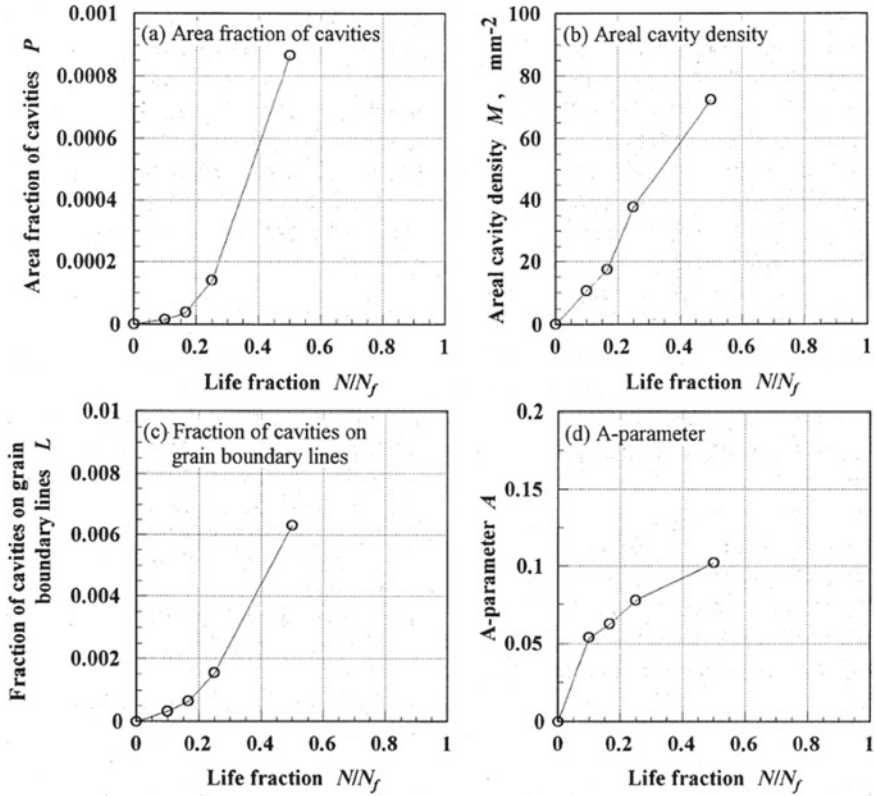


Fig. 2.19 Change in creep damage parameters with life fraction [11]

Note that $[\]_m$ in the above equation means that the values in parentheses are averaged over all grain boundaries. Additionally, using the average length of the grain boundary line l_{gm} , the area density of the grain boundary line n_g , and the average length of the cavity a_{cm} , the area density of the cavity M , the cavity occupancy on the grain boundary line L can also be expressed as follows.

$$L = \frac{a_{cm}}{l_{gm}} \cdot \frac{M}{n_g}. \quad (2.35)$$

Figure 2.19c shows the change over time of L . This parameter is analytically proven to be equal to the area fraction occupied by cavities on the grain boundary [16, 18]. Figure 2.19c shows that the value of L increases monotonically with increasing lifetime ratio and exceeds 0.6% around $1/2N_f$. This means that more than 0.6% of the grain interface is occupied by cavities at about half of the lifetime. This parameter has a clear physical meaning and corresponds to the effective cross-sectional area fraction used in conventional damage mechanics, making it a suitable parameter

for creep damage. However, further study is required to determine the relationship between the parameter values (limit values) when the material finally fractures, and damage caused by small cracks

The A-parameter, denoted as A , is defined as the percentage of grain boundary lines intersected by an inspection line oriented in the direction of the stress axis, where cavities are present. Thus, when the total number of grain boundary lines that intersect the inspection line is N_t and the number of those with cavities is N_c , the A-parameter A can be expressed by the following equation.

$$A = \frac{N_c}{N_t}. \quad (2.36)$$

Figure 2.19d shows the change over time of A . The rate of change of the A-parameter is high in the early stage of the lifetime, followed by a gradual decrease in the rate of increase. This is because the A-parameter does not consider the presence of multiple cavities on the same grain boundary as damage, even though multiple cavities are observed on the same grain boundary in the middle of the lifetime. Although this parameter is relatively easy to measure, its physical meaning is ambiguous, and the rate of increase tends to decrease in the mid-life, making it unsuitable for evaluating creep damage caused by intergranular cavities

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Chapter 3

Conditions for Appearance of Internal Intergranular Cracking Type Fracture Under Creep-Fatigue



Abstract This Chapter presents analytical results based on the experimental results and observations of cracks in creep-fatigue tests of a stainless steel SUS304. Two fracture morphologies were observed in high-temperature creep-fatigue fracture: inner cracking type fracture due to coalescence of small internal intergranular cracks and surface cracking type due to growth of small surface intergranular cracks. The fracture morphologies are strongly dependent on loading conditions. The inner cracking type fracture occurs at low tensile strain rate and high compressive strain rate, as well as at high temperatures, on the other hand, the surface cracking type fracture occurs under the opposite conditions. Three-dimensional fracture mechanism maps were created for tensile strain rate, compressive strain rate, and temperature to visualize the relationship between loading conditions and fracture morphologies. For the same strain range at high temperatures, the fatigue life at fatigue cracking type fracture is the longest, followed by the fatigue life at surface intergranular cracking type fracture and the fatigue life at internal intergranular cracking type fracture.

3.1 Introduction

As described in Chaps. 1 (Introduction) and 2, creep-fatigue failure of austenitic heat-resistant steels at high temperatures is dominated by the initiation and coalescence of intergranular cavities and the resulting initiation and propagation of small intergranular cracks [1–16]. Although small intergranular cracks can form inside the material [15, 16], in addition to the surface of the material [4–14]. Although the initiation and growth of small surface cracks and the resulting fracture (hereinafter referred to as surface cracking type fracture) have been clarified by experiments and analyses [4–14], there has been no systematic study of fracture caused by the initiation, growth, and coalescence of small inner cracks (hereinafter referred to as inner cracking type fracture), which is expected to appear under longer time and lower stress (or lower strain rate) conditions [4–14]. The difference between inner cracking type fracture and surface cracking type fracture has not been clarified either [17, 18]. Furthermore, in high-temperature creep-fatigue, the fracture mechanism

and morphology depend on the loading conditions, and also the life could depend on the fracture morphology. Critical factors that influence these conditions include tensile strain rate, compressive strain rate, strain range, and temperature.

In this chapter, creep-fatigue tests are conducted under various loading conditions (tensile strain rate, compressive strain rate, total strain range, and temperature) on a typical austenitic stainless steel SUS304, which is easily fractured at grain boundaries, to observe small cracks. The mechanism, morphology, and dependence on loading conditions of the inner cracking type fracture are clarified in comparison with surface cracking type fracture. Based on the experimental results, the creep-fatigue fracture is classified, the temperature and loading conditions that contribute to the shortest life are clarified, and a three-dimensional fracture mechanism map is created.

3.2 Test Methods

3.2.1 Specimen and Test Conditions

The specimens were solution-treated austenitic stainless steel SUS304 round bars, identical to those used in the previous studies [6, 8–14]. The chemical composition and heat treatment conditions are listed in Table 3.1. Figure 3.1 shows the shape and size of specimen. The specimens were solid round bar smooth specimen with a diameter of 10 mm and a parallel length of 20 mm. Before tests, the parallel parts were polished with emery paper up to #3000 to eliminate small surface flaws.

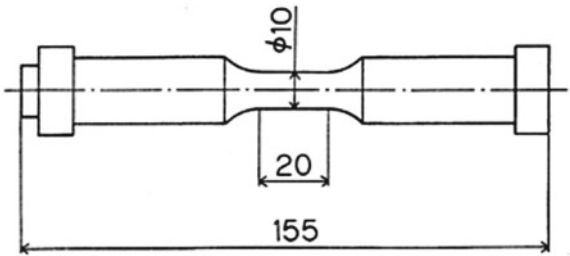
Fatigue tests were conducted under controlled tensile and compressive strains in air using an electric/hydraulic testing machine equipped with a high-frequency

Table 3.1 Chemical composition of specimen (wt.%) and condition of heat treatment

C	Si	Mn	P	S	Cu	Ni	Cr	Mo	Fe
0.06	0.31	1.07	0.31	0.018	0.44	8.68	18.37	0.25	Bal

Heat treatment: 1323K×1h → WQ

Fig. 3.1 Test specimen (dimension in mm)



induction heater. The test conditions are listed in Table 3.2. Cracks were predominantly observed in the specimens that had reached the failure life (stress reduced to three-fourths of the steady-state value). Since inner cracks are not affected by the atmosphere, whereas surface cracks are considered to be more affected, crack observations were performed on fractured specimens in vacuum (Test 10) [19] for surface cracking type fractures and the surface cracks were compared to the inner cracks.

3.2.2 Method of Observing Small Cracks

Figure 3.2 shows the observation areas of small cracks. Surface cracks were observed using an optical microscope and a scanning electron microscope after the oxide layer (approximately 2–3 μm) in area A on the specimen surface was removed by diamond paste polishing. For inner cracks, the specimens were split longitudinally, and the cross sections buffed, and area C was observed in the same procedure. The area of the observation region was 2.4 mm^2 in each case, and all cracks in the observation region were measured. The minimum crack length measured (half-length; hereafter all crack lengths are expressed as half-lengths) was 15 μm in each case.

The number, length, and tilt angle of small cracks in the observed region were measured by using photographs of regions A, B, and C magnified 200 times through an optical microscope. The crack length is the length projected perpendicular to the stress axis. In addition to the main crack, many cracks of different lengths are distributed in the failed specimen. Cracks which were initiated immediately after the coalescence of cavities on a single grain boundary or non-propagated cracks exist. Consequently, the tilt angle of a crack is defined as the angle between the straight line connecting the two ends of a crack of one grain boundary length and the direction perpendicular to the stress axis (Fig. 3.2).

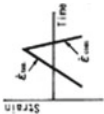
3.3 Results and Discussion

3.3.1 Comparison of Internal Intergranular Cracking Type Fracture and Surface Intergranular Cracking Type Fracture

3.3.1.1 Fracture Aspects

Figure 3.3 illustrates the small cracks in Test 1 with internal cracking type fracture and Test 9 and Test 10 with surface cracking type fracture (in vacuum). Both surface cracks and internal cracks initiated and grew at the grain boundaries.

Table 3.2 Test conditions and type of cracking [17]

Test no	Type of cracking*	Tensile strain rate $\dot{\epsilon}_{ten}$ (%/s)	Compressive strain rate $\dot{\epsilon}_{com}$	Total strain range $\Delta\epsilon_f$ (%)	Temperature T (K)	Environment	Failure life N_f (cycle)	Strain waveform
1	C _I	10 ⁻⁴	1	0.7	1073	Air	133	
2	C _I	10 ⁻³	1	0.7	1073	Air	255	
3	C _I	10 ⁻⁴	1	1.5	1073	Air	40	
4	C _I	10 ⁻³	1	1.5	1073	Air	76	
5	C _I	10 ⁻⁴	10 ⁻²	1.5	1073	Air	66	
6	C _I	10 ⁻⁴	1	1.5	1023	Air	80	
7	C _S	10 ⁻⁴	10 ⁻⁴	1.5	1073	Air	351	
8	C _S	10 ⁻²	1	1.5	1073	Air	163	
9	C _S	10 ⁻⁴	1	1.5	923	Air	180	
10	C _S	10 ⁻³	1	1.0	923	Vac	388	

*Note The letters “C_I” and “C_S” represent “Inner cracking type” and “Surface cracking type,” respectively

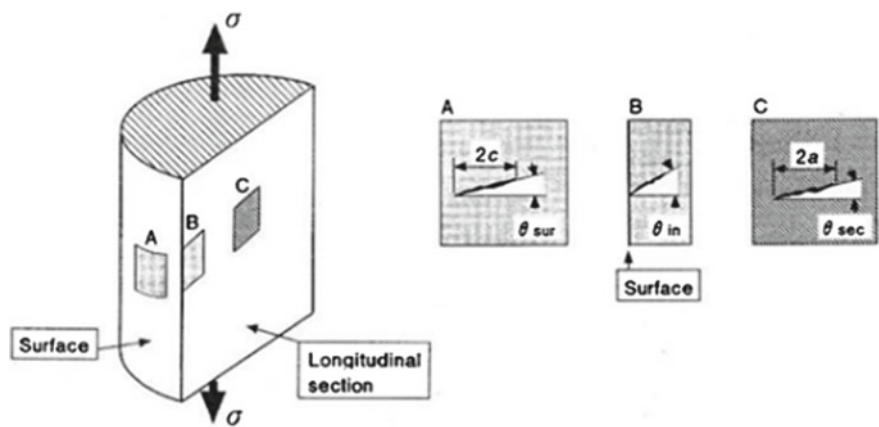


Fig. 3.2 Schematic illustration of crack observation areas [17]

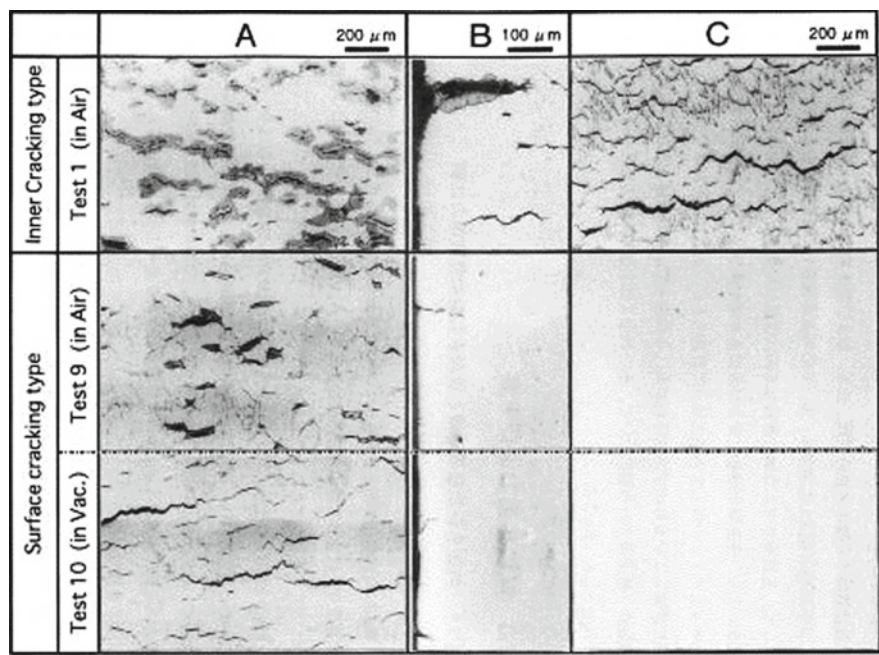


Fig. 3.3 Photographs of inner cracks and surface ones [17]

The photograph C in Fig. 3.3 reveals that there are no small cracks inside the specimen in Test 9 and Test 10, whereas numerous small cracks are observed in Test 1, indicating that the fracture morphologies of the two specimens are quite different. The comparison between Test 9 and Test 10 shows that the number of cracks differs

between the air and vacuum specimens, but both specimens have intergranular cracks only on the surface, indicating that the cracks are caused by creep.

Figure 3.4 shows the changes of the maximum and minimum stresses during the test for Test 1 (internal cracking type) and Test 10 (surface cracking type). The following distinct characteristics can be observed in the changes of the stresses for both cases: (1) In Test 10, the cyclic work hardening can be seen in the early stage of the test. On the other hand, in Test 1, the cyclic work hardening continued for a relatively long time and the stress gradually increased from the early to mid-life (approximately $0.50N_f$). The primary cause of this difference is the difference in test temperature. (2) In Test 10, the stress drops rapidly at the end of the lifetime due to the reduction of the cross-sectional area caused by the rapid growth of the main crack. On the other hand, in Test 1, the stress gradually decreases from the middle of the lifetime to failure. The characteristics of (1) and (2) suggest that under conditions in which a large number of small cracks are initiated in the specimen, the main crack does not form easily, and a long process of multiple small crack initiations is followed by rapid reduction in cross-sectional area due to the subsequent growth and coalescence of the cracks.

Fig. 3.4 Changes in maximum and minimum stresses. **a** Inner cracking type (Test 1). **b** Surface cracking type (Test 10) [17]

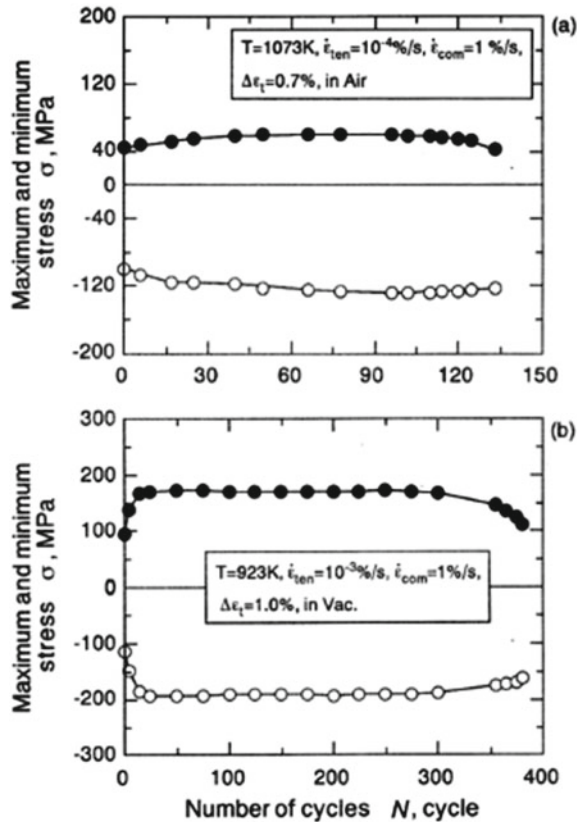
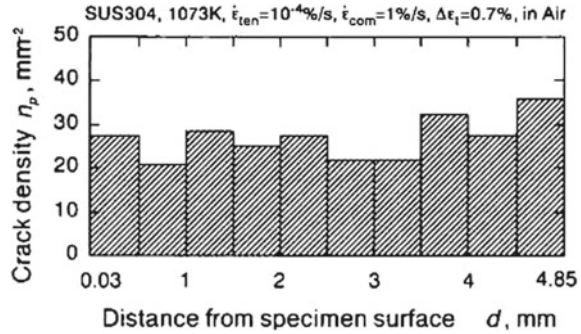


Fig. 3.5 Distribution of cracks on a longitudinal section inside the specimen (Test 1) [17]



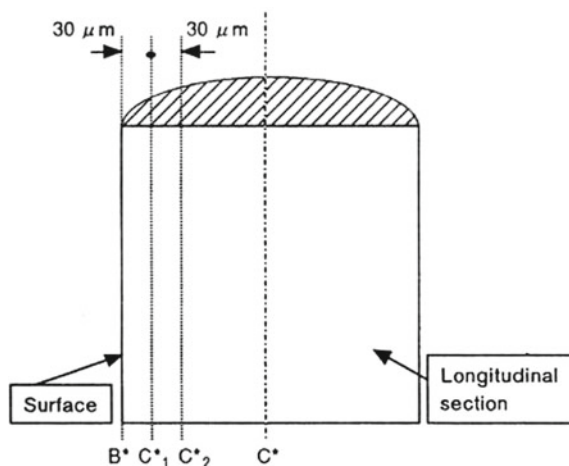
Subsequently, the distribution of the small inner crack density on the longitudinal section of the Test 1 specimen was examined. The number of cracks was measured from a distance of 30 μm from the specimen surface to the specimen center axis. The number of measurement planes was 19 in total on either side of the specimen center axis, and the crack observation area per plane was 1 mm².

The distribution of the measured crack density distribution is shown in Fig. 3.5. The horizontal axis denotes the distance from the specimen surface. The right end of the figure corresponds to the specimen center axis. As shown in the figure, there is no significant difference in the area density of small inner cracks, although there is some variation from region to region. In other words, the small inner cracks in this test are almost uniformly distributed up to the central axis, except in the near-surface region of the specimen where the oxide film is adhered (almost on the order of one grain boundary length). This result indicates that the specimen was subjected to uniform creep damage over the entire specimen under the test conditions.

The angular distribution of cracks on the longitudinal section in the Test 1 specimen was also examined. Cracks of one grain boundary size cut by a straight line parallel to the stress axis are only considered for measurement. As shown in Fig. 3.6, the cracks that appear on the longitudinal section are divided into three groups: the surface cracks in the depth direction (cracks cut on line B* in the figure), inner cracks not contacting the surface (cracks cut on lines C₁* and C₂*, approximately one grain boundary length (approximately 30 μm) and one grain size (approximately 60 μm) away from the surface in the figure) and inner cracks cut along the specimen center axis C*. The angles of 50 cracks cut in one straight line were measured.

Figure 3.7 illustrates the angular distribution of the measured cracks. The surface cracks cut on the B* line have an angular distribution centered at about 30°, while the cracks cut on the C*, C₁*, and C₂* lines are inclined perpendicular to the stress axis (0°) and have nearly identical angular distributions. In other words, the cracks inside the specimen at a distance of about one grain boundary length or more from the specimen surface have the same angular distribution without any surface effect. The difference in angular distributions between inner and surface small cracks will be discussed in the next section.

Fig. 3.6 Schematic illustration of crack observation sections [17]



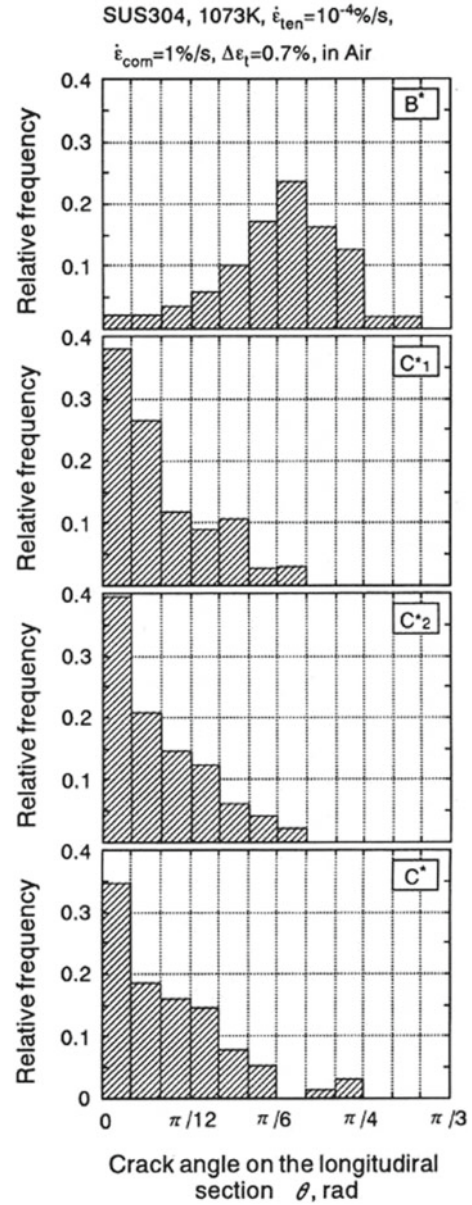
3.3.1.2 Crack Initiation Morphology

Figure 3.8a shows a scanning electron micrograph (SEM) of a typical small inner crack in the longitudinal section of Test 1 (region C in Fig. 3.2). Numerous cavities are observed on the grain boundary, and the cracks are formed by the interconnection of cavities. Figure 3.8b shows a photograph of a surface small crack on the surface of Test 10 (region A in Fig. 3.2). A grain boundary wedge-shaped crack is observed, demonstrating a morphological distinction from that of inner small crack.

Figure 3.9 shows SEM photographs of the fracture surfaces of Test 1 and Test 10 specimens that were broken using tension at low temperatures (liquid nitrogen temperature) after interruption at 75% of life. Both specimens failed by intergranular fracture, but the fracture surface of Test 1 (Fig. 3.9a–c), which is of the inner cracking type, has numerous spherical dimples on the grain boundary facets, including precipitates, likely carbides. The absence of these dimples on the fracture surface of Test 10 (Fig. 3.9b) and their shapes and sizes of these dimples are almost identical to those of the cavities on the longitudinal section shown in Fig. 3.8a, suggesting that these cavities correspond to creep cavities initiated and grown on the grain boundary with carbides as their nuclei. Figure 3.8a.

Figure 3.10 shows the angular distributions of surface and inner cracks in Test 1 and Test 10 relative to the vertical direction of the stress axis. For an isotropic polycrystalline material such as this material, the grain boundary inclination in regions A and C is uniform in all directions. The depth distribution of the surface grain boundaries in region B is not uniform, and the probability of the grain boundary being close to the perpendicular direction of the stress axis is higher [20]. Therefore, the angular distribution of cracks observed in region B does not directly correspond to the probability of crack initiation at each angle, nor can it be directly compared with the results measured in regions A and C. Considering the fact that the probability density distribution of the angles of surface grain boundaries in the depth direction

Fig. 3.7 Distribution of cracking angles on a longitudinal section of the specimen (Test 1) [17]



is given by $\cos \theta_{in}$ [20], the relative frequency distribution of the angles for small cracks in region B is divided by the cosine of the median value of each angle rank, $\cos \theta_{in}$.

The results for region C in Fig. 3.10 show that inner cracks tend to initiate on the grain boundary facets near perpendicular to the stress axis. In other words, the

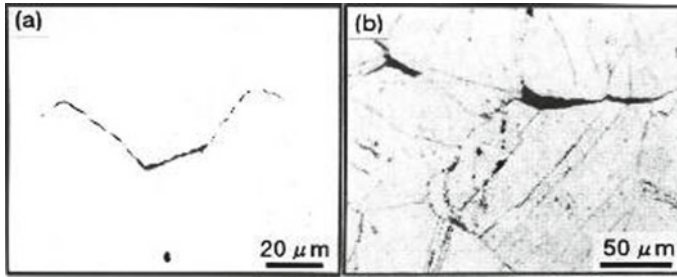


Fig. 3.8 Scanning electron micrographs of intergranular cracks. **a** Inner cracking type observed on a longitudinal section (Test 1). **b** Surface cracking type observed on a surface (Test 10) [17]

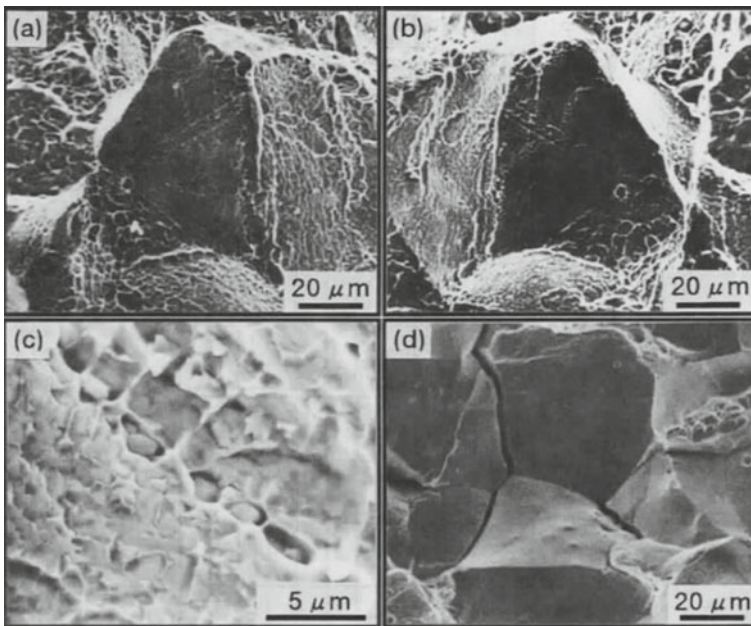


Fig. 3.9 Scanning electron micrographs of the fracture surface which were obtained by brittle fracture at 200 K after creep-fatigue tests ($3/4N_f$). **a** Inner cracking type (Test 1). **b** The other fracture surface matching with (a). **c** Magnification of (a). **d** Surface cracking type (Test 10) [17]

normal stress on the grain boundary facets dominates crack initiation. It has been reported that the diffusion of atomic vacancies becomes active under creep conditions under low stress (or low strain rate) at high temperatures, and that the normal stress on the grain boundary induces diffusion of vacancies along the grain boundary and cavity initiation and growth on the grain boundaries[21–23]. Figures 3.8, 3.9, and 3.10 indicate that this diffusion-dominated cavity growth and coalescence is the mechanism of small inner crack initiation in high-temperature creep-fatigue.

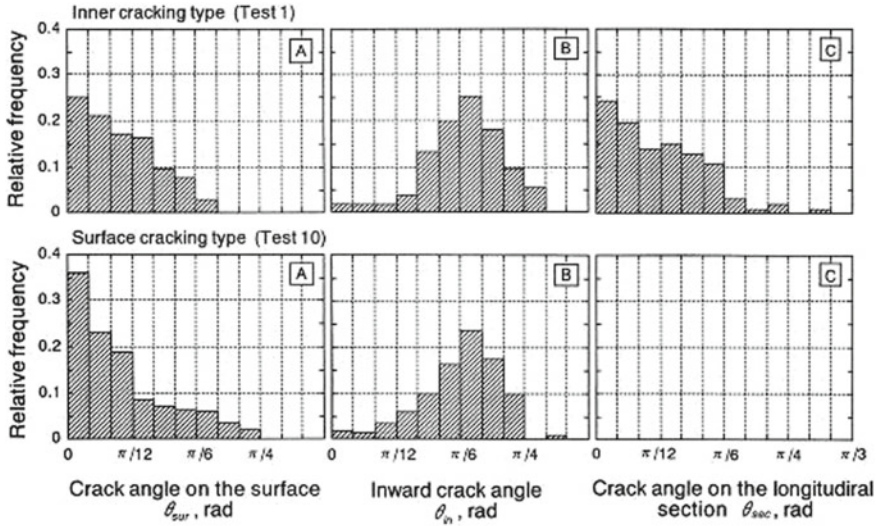


Fig. 3.10 Distribution of cracking angles in the domains A, B, and C. The upper figures show the results of inner cracking type (Test 1) and the lower ones surface cracking type (Test 10) [17]

On the other hand, Fig. 3.10 illustrates that the surface cracks are almost perpendicular to the stress axis on the surface (region A), while the largest number of surface cracks in the depth direction (region B) is nearly 35° . This suggests that the intergranular slip in surface direction predominantly influences to the crack initiation. The amount of intergranular slip inside the material is small because the deformation that mismatches caused by such slip is not easily absorbed. In contrast, the intergranular slip on surface, being less constrained, is more likely to initiate. This may be the reason why the intergranular cracks are initiated only on the surface during creep-fatigue at medium temperatures and high stress.

3.3.1.3 Fracture-Inducing Main Crack Formation

Figure 3.11 illustrates the fracture-induced main crack in the longitudinal section. In Test 1, a main crack with a length of approximately 1.5 mm is observed, as shown in Fig. 3.11a. This crack is located in the center of the specimen and does not touch the surface. In the case of an inner cracking type fracture, the main crack consists of several sections with different openings. Given the high crack density, this suggests that many small cracks are connected to a long crack at the end of the life. On the other hand, in the case of Test 10, as shown in Fig. 3.11b, a main crack with a length of approximately 4.5 mm on the surface, resulting in failure due to this surface crack.

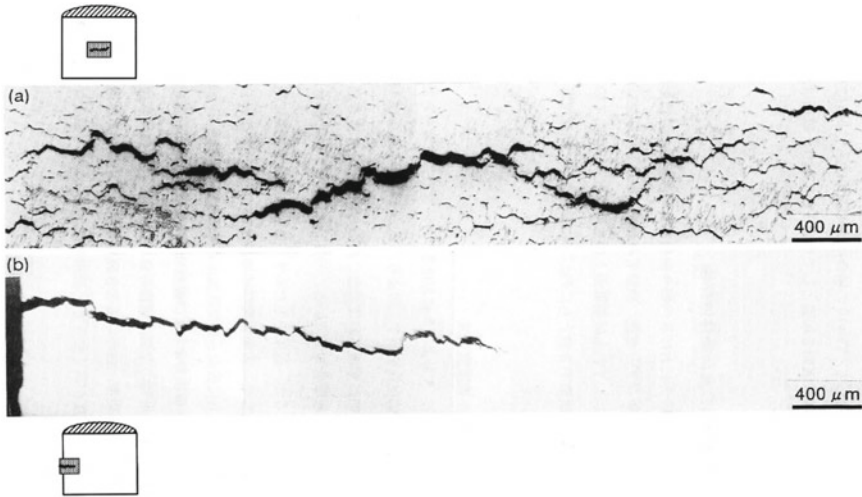


Fig. 3.11 Morphology of main cracks. **a** Inner cracking type (Test 1). **b** Surface cracking type (Test 10) [17]

3.3.2 Dependence of Fracture Morphology on Loading Conditions

Next, the dependence of the inner cracking type and surface cracking type fracture morphology on loading conditions (tensile strain rate, compressive strain rate, total strain range, and temperature) in creep-fatigue is examined.

The photographs shown below are all taken on longitudinally split cross sections, and both small inner cracks and small surface cracks initiated and grew at the grain boundaries as shown in Fig. 3.3 for all test conditions.

3.3.2.1 Tensile Strain Rate Dependence

Figure 3.12 shows the tensile strain rate dependence of the main crack morphology that caused creep-fatigue failure in Tests 3, 4, and 8. The figures show that the main cracks formed inside the specimen at $\dot{\epsilon}_{\text{ten}} = 10^{-4}\%/s$ and $\dot{\epsilon}_{\text{ten}} = 10^{-3}\%/s$, and that the coalescence of numerous small internal intergranular cracks was dominant, indicating an inner cracking type fracture. On the other hand, at $\dot{\epsilon}_{\text{ten}} = 10^{-2}\%/s$, although some small inner cracks are observed, the main cracks that cause failure are positioned on the surface and grow while absorbing the inner cracks, resulting in a surface cracking type fracture. The specimen tends to transition from surface cracking type fracture to inner cracking type fracture as the tensile strain rate decreases.

Figure 3.13 shows the relationship between fatigue life, inner crack density, and tensile strain rate. The vertical axis's left segment represents fatigue life (denoted

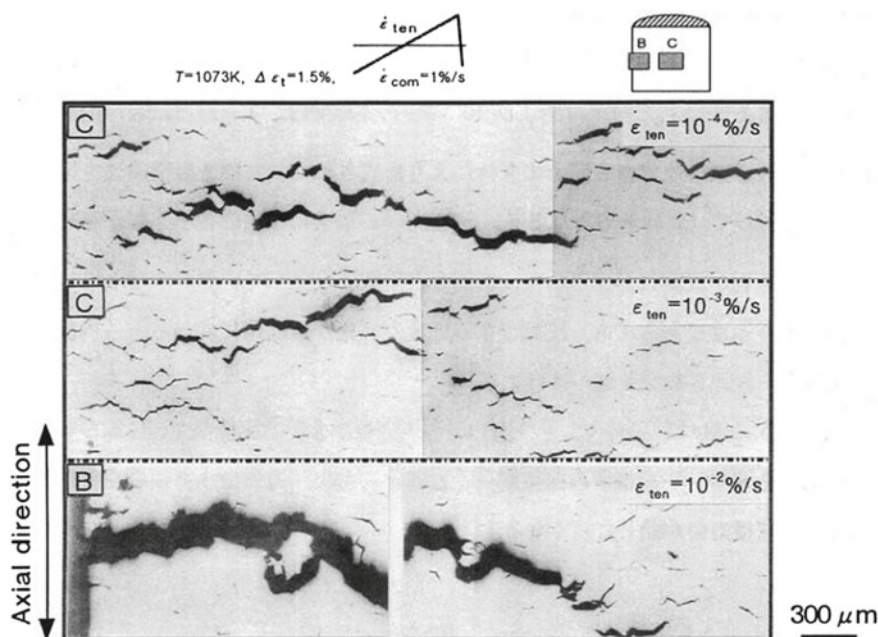


Fig. 3.12 Micrographs showing the dependence of crack morphology on tensile strain rate [17]

by squares), while its right counterpart portrays inner crack density (indicated by circles). The black and white marks indicate the inner cracking type and surface cracking type, respectively. A significant dependency on the tensile strain rate is conformed, with lower tensile strain rates resulting in higher inner crack density and shorter fatigue life.

3.3.2.2 Compressive Strain Rate Dependence

Figure 3.14 shows the compressive strain rate dependence of the main crack morphology in Tests 3, 5, and 7. The figures show that for $\dot{\epsilon}_{\text{com}} = 1\%/s$ and $10^{-2}\%/s$, the main cracks are formed by the coalescence of many small intergranular cracks distributed inside the specimen, resulting in an inner cracking type fracture, whereas for the symmetric waveform of $\dot{\epsilon}_{\text{com}} = 10^{-4}\%/s$, no small inner cracks are observed and the main cracks are formed on the surface, resulting in a surface cracking type fracture.

Figure 3.15 shows the relationship between fatigue life, inner crack density, and compressive strain rate. As the compressive strain rate increases, the inner crack density increases and the fatigue life becomes significantly shorter.

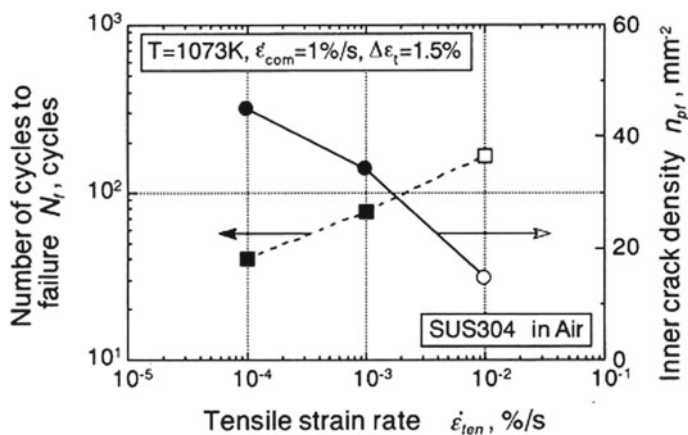


Fig. 3.13 Dependence of fatigue life and inner crack density on tensile strain rate. The inner crack density, n_{pf} was measured at the center of a longitudinal section. Solid marks for inner cracking type and hollow marks for surface cracking type [17]

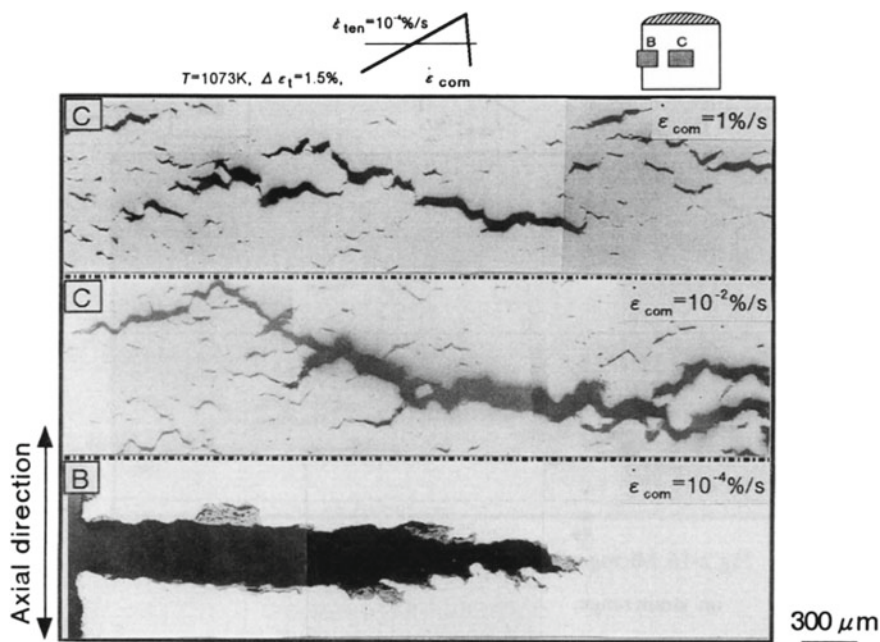


Fig. 3.14 Micrographs showing the dependence of crack morphology on compressive strain rate [17]

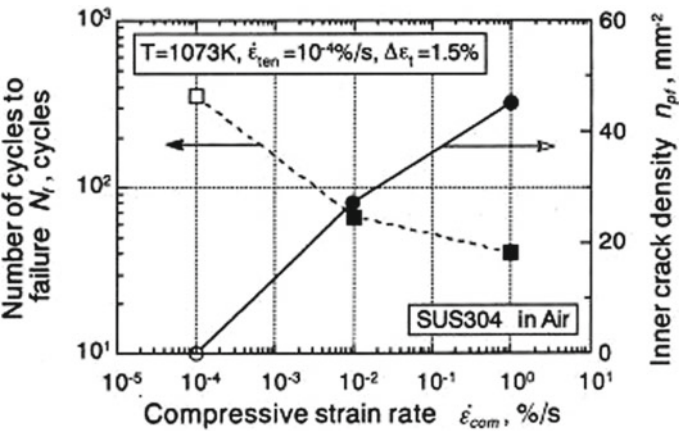


Fig. 3.15 Dependence of fatigue life and inner crack density on compressive strain rate. The inner crack density, n_{pf} was measured at the center of a longitudinal section. Solid marks for inner cracking type and hollow marks for surface cracking type [17]

3.3.2.3 Dependence on Total Strain Range

Figure 3.16 shows the dependence of the main crack morphology on the total strain range for Tests 3 and 1. The figure shows that many small inner cracks are observed for both $\Delta\epsilon_t = 1.5\%$ and 0.7% , and the main cracks are formed inside the material, indicating an inner cracking type fracture. The fracture morphology is less dependent on the total strain range.

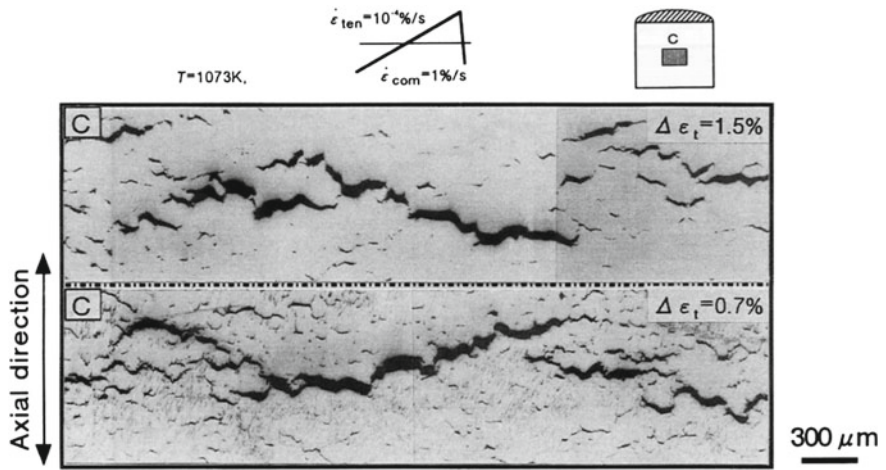


Fig. 3.16 Micrographs showing the dependence of crack morphology on strain range [17]

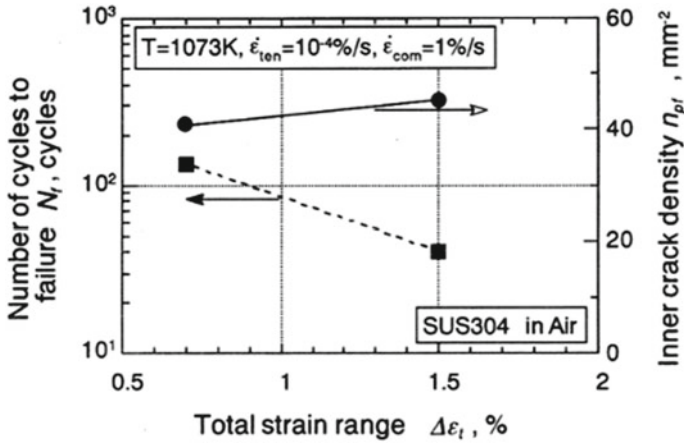


Fig. 3.17 Dependence of fatigue life and inner crack density on strain range. The inner crack density, n_{pf} was measured at the center of a longitudinal section. Solid marks for inner cracking type [17]

Figure 3.17 shows the relationship between fatigue life, inner crack density, and total strain range. As the total strain range increases, inner crack density increases and the fatigue life decreases.

3.3.2.4 Temperature Dependence

Figure 3.18 shows the temperature dependence of the main crack morphology in Tests 3, 6, and 9. The figures show that at $T = 1073$ K and 1023 K, the main cracks formed inside the specimen, indicating an inner cracking type fracture. On the other hand, at $T = 923$ K, no small cracks are observed except on the surface, and the main cracks form on the specimen surface, indicating a surface cracking type fracture. As the temperature increases, the fracture morphology shifts from a surface cracking type fracture to an inner cracking type fracture.

Figure 3.19 shows the relationship between fatigue life, inner crack density, and temperature. As the temperature increases, the inner crack density increases and the fatigue life is significantly reduced.

3.3.3 Dependence of the Average Increase Rate of Crack Density on Loading Conditions

The frequency of crack initiation per strain cycle is used as an indicator of the difficulty of crack initiation. For surface cracks, it is the increase rate of the crack density (number of cracks per unit area) on the specimen surface. The average increase rate

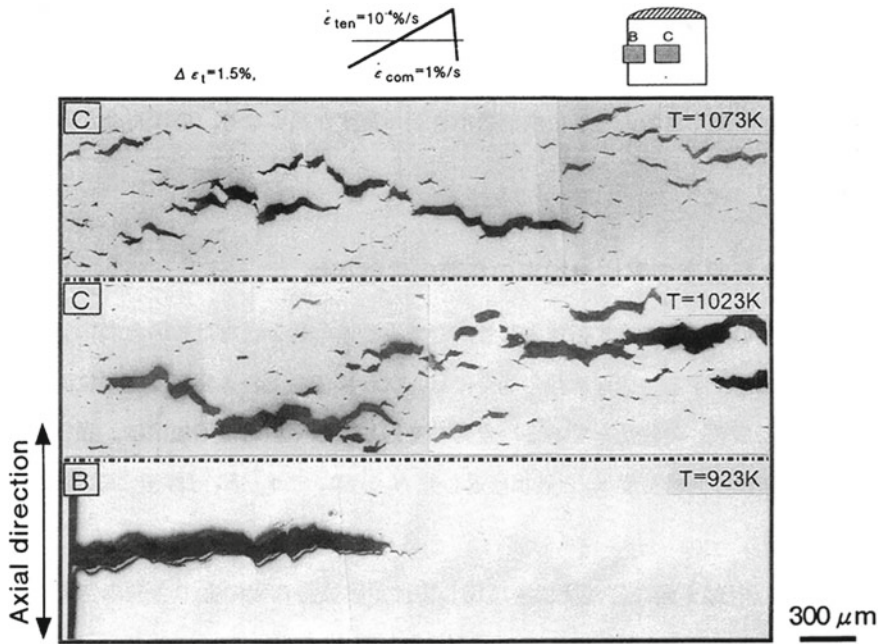


Fig. 3.18 Micrographs showing the dependence of crack morphology on temperature [17]

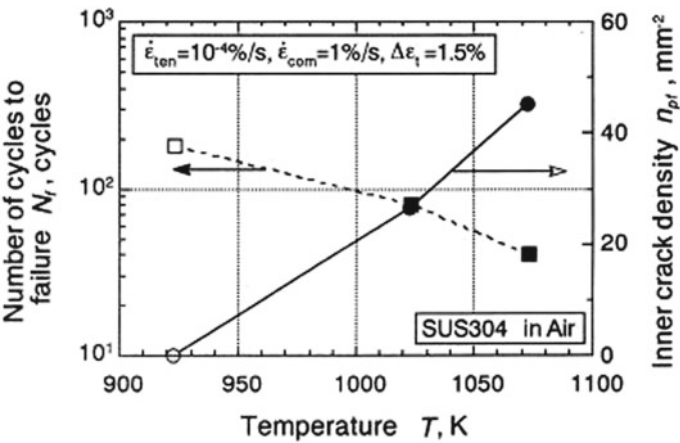


Fig. 3.19 Dependence of fatigue life and inner crack density on temperature. The inner crack density, n_{pf} , was measured at the center of a longitudinal section. Solid marks for inner cracking type and hollow marks for surface cracking type [17]

over the entire life can be evaluated as n_{sf}/N_f using the area density of small cracks, n_{sf} , and the number of cycles to failure, N_f .

Similarly, the increase rate of the inner crack density can be evaluated as n_{vf}/N_f using the spatial density of small cracks (number of cracks per unit volume) n_{vf} and the number of cycles to failure N_f . Information on inner cracks can be obtained from observations on a longitudinal section parallel to the stress axis. However, the area density of cracks in a longitudinal section does not directly reflect the spatial density of small cracks in the specimen. Therefore, the volume density of the inner crack is estimated as follows. Assuming that the inner crack shape is disk-shaped, the relationship between the volume density n_v and the area density n_p of the longitudinal section at a certain number of cycles can be expressed as follows from probability theory [24].

$$n_v = n_p / (2r_m), \quad (3.1)$$

where r_m is the average radius of the inner crack. The above equation indicates that the probability of a crack appearing in the longitudinal section increases as the radius of the inner crack increases, and the number of cracks in the section increases as the mean radius r_m increases, even though the volume density of the inner crack is constant. On the other hand, when all the inner cracks have the same radius r_a , r_a can be obtained from the average length (half-length) of the cracks on the section as a_m by the following equation.

$$r_a = 4a_m / \pi \quad (3.2)$$

Here, the average radius of the inner cracks is expressed by the following equation.

$$r_m = 4a_m / \pi \quad (3.3)$$

Substituting (3.3) into (3.1), we obtain the following equation.

$$n_v = n_p \times \pi / (8a_m). \quad (3.4)$$

The volume density of the inner crack can be estimated from the equation. Here, n_{pf} and a_{mf} are obtained from the observation of cracks on the longitudinal section of the fractured specimen after the test, and the volume density n_{vf} is determined from (3.4). The inverse analytical method [24] proposed by Tada et al. can be used to perform a precise analysis considering the random distribution of the inner crack radius. Equation (3.3) is found to be a reasonable approximation.

Figure 3.20 shows the average increasing rate of crack density n_{sf}/N_f for surface and inner cracks and the dependence of n_{vf}/N_f on tensile strain rate, compressive strain rate, total strain range, and temperature. Both inner and surface cracks show similar trends for each condition, with (1) a slower tensile strain rate, (2) a higher compressive strain rate, (3) a larger total strain range, and (4) a higher temperature resulting in a larger increase in the increasing rate and crack initiation. Although it

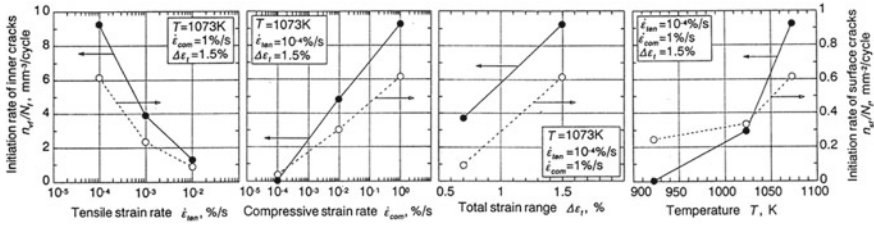


Fig. 3.20 Dependence of average increasing rate of crack density (i.e., average crack initiation rate) on loading conditions [17]

is not possible to directly compare the density and growth rate of two-dimensional (planar) surface cracks and three-dimensional (spatial) inner cracks, the dependence of inner cracks on loading conditions is greater than that of surface cracks. This causes the fracture morphology shift from a surface cracking type fracture to an inner cracking type fracture, or vice versa, depending on the loading conditions.

The average crack length c_{mf} (surface crack) or average radius r_{mf} (internal crack) divided by the failure life N_f is taken as a parameter representative of the average crack propagation rate over the entire life. The r_{mf} value is calculated from (3.3). Figure 3.21 shows the dependence of c_{mf}/N_f and r_{mf}/N_f on loading conditions. The inner and surface cracks have similar trends for each loading condition, with the average crack length increasing at (1) slower tensile strain rates, (2) higher compressive strain rates, (3) larger total strain ranges, and (4) higher temperatures. These trends are similar to the loading condition dependence of the crack density increasing rate shown in Fig. 3.20. The average length increasing rate for inner cracks has a greater dependence on each test condition than that for surface cracks. However, since inner cracks can coalesce with neighboring cracks in any direction, it is possible that they grow with more mergers and that the apparent increasing rate is greater for inner cracks than for a single crack.

Figure 3.22 shows the creep-fatigue fracture conditions based on the observations of small cracks and main cracks in this test. The vertical axis of the figure shows the crack density on the longitudinal section of the fractured specimen, n_{pf} , multiplied by the average crack length, a_{mf} . Using $n_p = 8/\pi \cdot a_m \cdot n_v$ from (3.4) and $a_m = \pi/4 \cdot r_m$ from (3.3), the following equation is obtained.

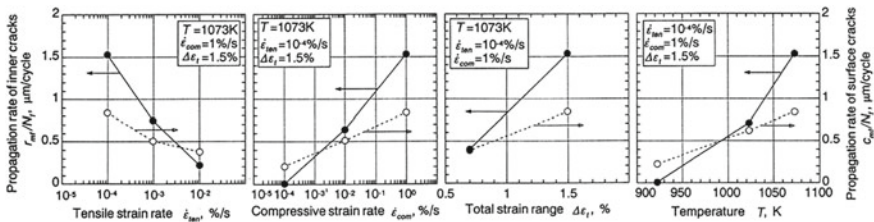
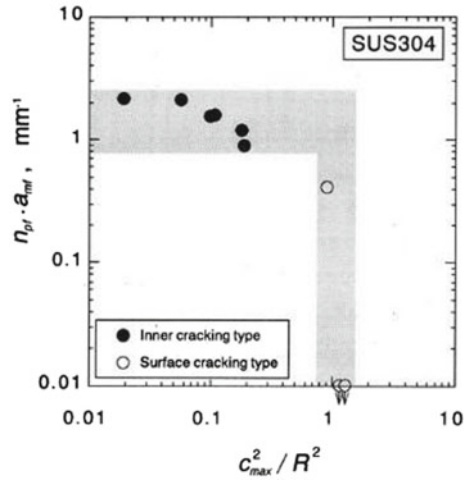


Fig. 3.21 Dependence of average crack propagation rate on loading conditions [17]

Fig. 3.22 Failure condition in inner cracking type and surface cracking type fracture [17]. Ordinate, $n_{pf} \cdot a_{mf}$, represents total area of inner cracks involved in a unit volume and abscissa, $(c_{\max}/R)^2$, total area of surface cracks involved in a unit cross-sectional area



$$n_p \cdot a_m = \pi/2 \cdot n_v \cdot r_m^2 \quad (3.5)$$

It represents the product of the volume density of the inner crack and the square of the mean radius, i.e., a value proportional to the total area of the inner crack per unit volume. On the other hand, the horizontal axis in the figure is the maximum crack length on the surface of the fractured specimen, $c_{\max}$ squared, divided by the specimen radius, R squared. This is proportional to the crack area per unit transverse sectional area.

The figure shows that the crack area per unit volume is a constant fracture condition for the inner cracking type fracture. This is because once the crack area per unit volume reaches a certain value, the main crack is formed by the rapid coalescence of small cracks. On the other hand, the surface cracking type fracture indicates that one or several large surface cracks, which also grow in the depth direction, cause failure when they reach a certain length.

3.3.4 Classification of High-Temperature Fatigue Fracture Types and Fracture Mechanism

3.3.4.1 Classification of High-Temperature Fatigue Fracture Modes

In high-temperature fatigue with high loading rate, it is known that surface cracks initiated by crystallographic slip, similar to that in room temperature fatigue, grow within the grain and cause failure [2, 3]. This phenomenon is referred to as pure fatigue-type transgranular fracture. In contrast, the surface crack observed in this study is caused by intergranular slip due to in-grain dislocation creep, and its growth

path is also intergranular. The other inner crack is the same creep-dominated crack, but the main cause is considered to be the initiation, growth, and coalescence of intergranular cavities due to intergranular diffusion creep. Based on the above, high-temperature fatigue fracture can be roughly classified into (1) internal intergranular cracking type fracture (C_I), (2) surface intergranular cracking type fracture (C_S), and (3) pure fatigue-type transgranular fracture (F_S). Figure 3.23 shows the typical cracking behavior of the above three fracture types. The F_S photo was observed using a specimen from a previous fatigue test of the same material at 923 K in a vacuum.

Figure 3.24 [14, 25] shows the fracture morphology and process of high-temperature fatigue. C_I is an intergranular diffusion-dominated inner cracking type fracture similar to long-term creep at high temperatures, very-slow-tensile fast-compressive conditions and C_S is an intergranular slip-dominated fracture at medium temperatures, medium-speed-tensile fast-compressive or slow-tensile fast-compressive conditions. F_S is a pure fatigue-type transgranular fracture similar to the room temperature fatigue that occurs at low-temperature, fast-tensile, or fast-compressive conditions. As described below, the failure lives of ductile steels such as Type 304 tend to increase in this order.

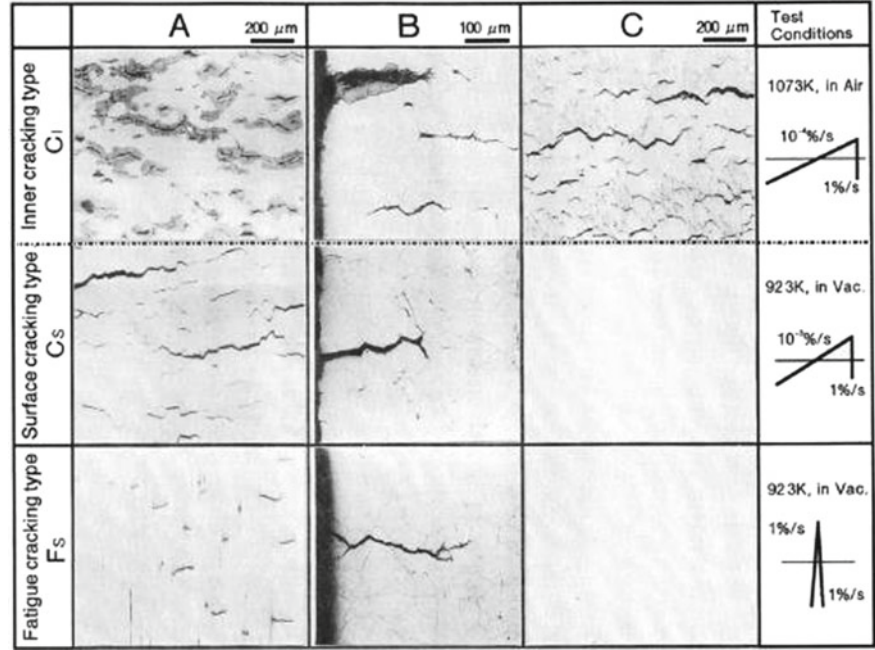


Fig. 3.23 Photographs of high-temperature fatigue cracks are classified into three types. Type C_I , inner cracking type; Type C_S , surface cracking type; Type F_S , fatigue cracking type [17]

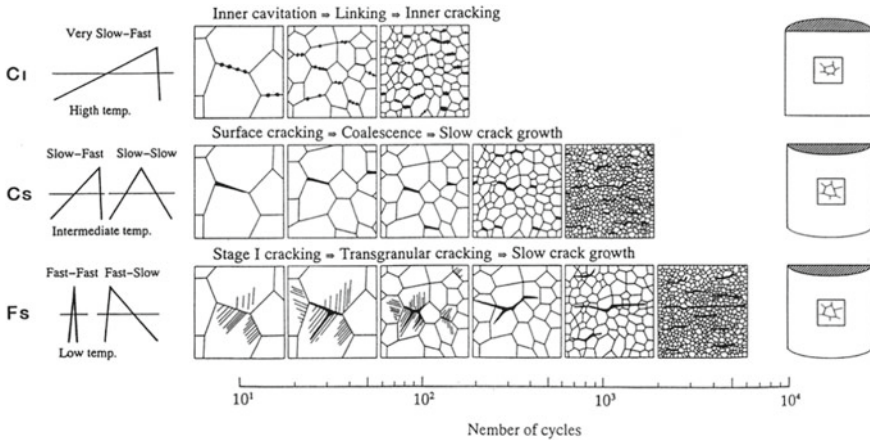


Fig. 3.24 Classification of the type of fatigue fracture in ductile polycrystalline alloys at high temperature: Type C_I , inner cracking type; Type C_S , surface cracking type; Type F_S ; fatigue cracking type [17]

3.3.4.2 High-Temperature Fatigue Fracture Mechanism

Figure 3.25 shows the regions where the above three types of fracture types appear for Type 304. As previously highlighted, the differences in fracture modes are considered to be caused by the difference in fracture mechanisms, and this figure can be referred to as a fracture mechanism map of creep-fatigue. Incorporating the conventional test results of Ohtani et al. [26], which include the findings from this study, this figure also integrates the results of common tests conducted by the Iron and Steel Institute of Japan [27]. The results are categorized into pure fatigue-type transgranular fracture (marked with ■), surface intergranular cracking type fracture (marked with ○), and internal intergranular cracking type fracture (marked with ●), corresponding to the test conditions (temperature, tensile strain rate, compressive strain rate). As illustrated in the figure, as (1) the tensile strain rate decreases, (2) the compressive strain rate increases, and (3) the test temperature increases, the fracture modes of high-temperature fatigue shift from fatigue-dominated to creep-dominated in the following order: (a) pure fatigue-type transgranular fracture (fatigue-dominated) → (b) surface intergranular cracking type fracture (b) surface intergranular cracking type fracture (creep-dominated) → (c) internal intergranular cracking type fracture (creep-dominated).

Figure 3.26 shows the relationship between the inelastic strain range and the fatigue life. The figure also shows the results of previous tests[26, 28], which are classified into ■ (pure fatigue-type transgranular fracture), ○ (surface intergranular cracking type fracture), and ● (internal intergranular cracking type fracture), corresponding to Fig. 3.25. The lower limit line in the figure is obtained by integrating the following Monkman–Grant relationship [28] obtained from static creep tests over one cycle, τ , and converting it to the relationship between the inelastic strain range,

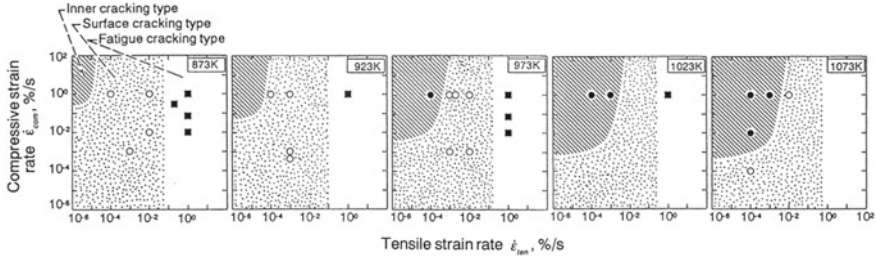


Fig. 3.25 Creep-fatigue fracture mechanism maps [17]

$\Delta\epsilon_{in}$, and the number of cycles to failure, N_f , as in (3.7).

$$\dot{\epsilon}_c \cdot t_r = \text{Const} \quad (3.6)$$

$\dot{\epsilon}_c$: steady-state strain rate, t_r : rupture time, integration over one period τ

$$\dot{\epsilon}_c \tau = \Delta\epsilon_{in}, \quad t_r = N_f \cdot \tau. \quad (3.7)$$

The upper line is the Manson–Coffin law [29, 30] obtained from pure fatigue tests. The fatigue life of the inner cracking type is close to that of static creep. Considering this and the failure mode, it can be inferred that the damage and fracture mechanisms of the inner cracking type are almost the same as those of static creep. Even though the macroscopic strain is repeated within a certain range, the creep deformation in tension and plastic deformation in compression are not balanced locally near the grain boundary, and the tensile creep deformation accumulates in one direction as in the static creep case, resulting in crack initiation and growth.

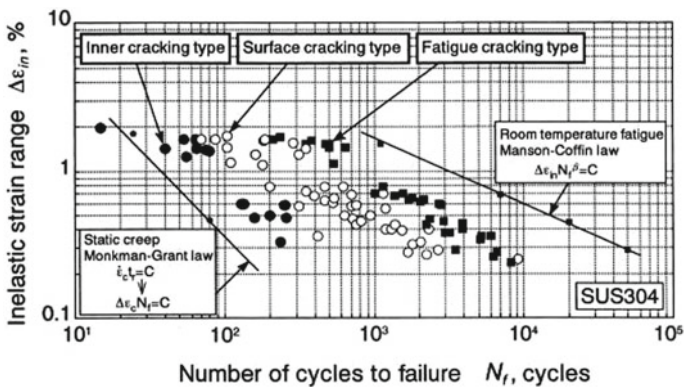


Fig. 3.26 Dominated fracture mode (cracking type) in the relationship between inelastic strain range and fatigue life [17]

The figure shows that there is a large difference (variation) in fatigue life even for the same intergranular fracture. One of the reasons for this is that the mechanism and morphology of fracture change with cyclic loading conditions such as strain rate in high-temperature creep-fatigue as mentioned above. As shown in the figure, the fatigue lives for the same inelastic strain range are shortened in the order of (a) pure fatigue-type transgranular fracture, (b) surface intergranular cracking type fracture, and (c) internal intergranular cracking type fracture. From this, we can deduce that the most dangerous creep-fatigue fracture morphology of heat-resistant steel is that caused by the initiation, growth, and coalescence of small intergranular cracks inside the material.

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Chapter 4

Initiation and Growth Behavior of Small Inner Crack



Abstract In this chapter, using austenitic stainless steel SUS304 and steam turbine rotor material 1Cr-1Mo-1/4 V forged steel, creep-fatigue interruption tests were conducted at high temperatures and low tensile strain rates. The small cracks that initiated and grew inside the specimens were observed on the longitudinal split sections along the stress axis, and their changes with time were clarified. In addition, the size of intergranular cavities, which are considered to be closely related to the inner cracking type creep-fatigue fracture, was measured. The main results obtained can be summarized as follows. Small inner cracks are initiated and grow with a temporal and spatial distribution on grain boundaries nearly perpendicular to the stress axis. However, there are few cases in which neighboring cracks coalesce more than one grain size (about 50 μm) distance apart in the direction of the stress axis. Small inner cracks initiate in the early stage of the lifetime. The density of the cracks increases rapidly after the mid-life. Three-dimensional observation of a $0.75N_f$ specimen of SUS304 shows that the areal crack density of the longitudinally split section is about 35 cracks/ mm^2 at this point, but the volumetric crack density is approximately 1400 cracks/ mm^3 , which is very high. Multiple cavities are initiated and grow in a series on the same grain boundary, and they merge to form a crack with the length of a grain boundary as a unit. The difficulty of crack initiation is almost independent of the grain boundary length, but cracks tend to be initiated at grain boundaries that are perpendicular to the stress axis. Crack initiation is dominant until late in the fatigue life. Small inner crack growth also progresses discretely with one grain boundary length as a unit due to the growth and coalescence of cavities on almost one grain boundary adjacent to the crack tip and is similar to crack initiation in mechanism. However, there is no significant difference in the crack length distribution up to the late fatigue life, and no significant crack growth is observed. At the late stage of fatigue life, the overall crack length distribution rapidly shifts toward the longer side, suggesting that a large number of small cracks distributed on almost the same cross section grow slightly to induce mutual coalescence and rapidly form a main crack.

4.1 Introduction

In this chapter, among the high-temperature fatigue fractures of heat-resistant steels, the internal intergranular cracking type fracture caused by the growth and coalescence of numerous small cracks that uniformly initiate inside the material under high temperature and low tensile strain rate (or low stress) conditions is extremely dangerous because it leads to the shortest fatigue life. Additionally, the formation mechanism of small inner cracks is different from that of small surface cracks, with the growth and coalescence of creep cavities being dominant, as pointed out in Chap. 2. Considering the fact that creep-fatigue strength under high temperature, low stress, and long period loading conditions is an issue in actual high-temperature equipment used for long periods of time [1–4], it is important to understand the initiation and growth behavior of small cracks inside materials. However, there have been no systematic studies on the changes in the initiation and growth of small inner cracks over time. For inner cracks, it is impossible to directly and continuously observe the initiation and growth process of the same crack within a specific region. This makes the characterization of inner cracks very difficult.

In this chapter, creep-fatigue interruption tests were conducted on austenitic stainless steel SUS304 and 1Cr-1Mo-1/4 V forged steel, which are typical heat-resistant steels, under conditions that exhibit internal cracking type fracture. The damage caused by numerous cavities in the specimen and the fracture process leading to the initiation and growth of small inner cracks are discussed.

4.2 Test Methods

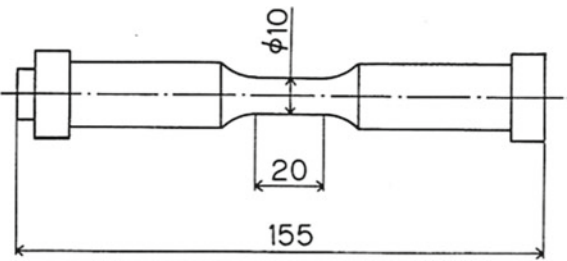
4.2.1 Materials and Specimens

Two types of materials were used: solution-treated austenitic stainless steel SUS304 round bars which is the same as used in Chap. 3 and 1Cr-1Mo-1/4V forged steel cut from unused turbine rotors. Table 4.1 lists the chemical composition and heat treatment conditions of the materials. The specimens used for both steels were solid round smooth specimens with a diameter of 10 mm and a parallel length of 20 mm, as shown in Fig. 4.1. Before the test, the parallel parts were polished with emery paper up to No. 3000 to remove small surface flaws.

4.2.2 Test Conditions

The creep-fatigue tests were conducted using an electro-hydraulic servo fatigue testing machine equipped with a high-frequency induction heater, under atmospheric conditions, using the asymmetric triangular waveform (cp-type) of low-speed tension

Fig. 4.1 Test specimen
(dimensions in mm)



and high-speed compression with strain control shown in Fig. 4.2. The detailed test conditions are presented in Table 4.2. The failure lives (the number of cycles when the tensile stress decreases to 3/4 of the steady state) N_f of the tests are 133 cycles (SUS304) and 86 cycles (1Cr-1Mo-1/4 V steel), respectively. In order to observe the change over time of small cracks inside the specimens, the interrupted specimens were tested at 10% (13 cycles), 17% (22 cycles), 25% (33 cycles), 50% (66 cycles), 75% (100 cycles), and 100% (133 cycles) of the failure life for SUS304 under the same test conditions. The 1Cr-1Mo-1/4 V steel specimens were subjected to 25% (21 cycles), 50% (43 cycles), 75% (65 cycles), and 100% (86 cycles) of the failure life. For convenience, these specimens will be referred to as $0.10N_f$, $0.17N_f$, $0.25N_f$, $0.50N_f$, $0.75N_f$, and N_f specimens, respectively.

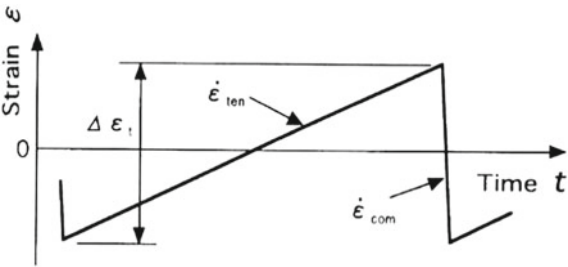


Fig. 4.2 Strain waveform

Table 4.2 Test conditions

Material	Tensile strain rate $\dot{\epsilon}_{ten}$ (%/s)	Compressive strain rate $\dot{\epsilon}_{com}$ (%/s)	Total strain range $\Delta \epsilon_t$ (%)	Temperature T (K)	Failure life N_f (cycle)
SUS304	1×10^{-4}	1	0.7	1073	133
1Cr-1Mo-1/4 V	3×10^{-4}	1	1.0	923	86

4.2.3 Methods of Observing Small Cracks

4.2.3.1 Two-Dimensional Observation

The parallel parts of the interrupted and fractured specimens were split longitudinally in the direction of the load axis, and the cross sections were polished with emery paper up to No. 2000 and then buffed and electropolished for SUS304 and 1Cr-1Mo-1/4 V steel, respectively. The small cracks were observed and measured by an optical microscope and scanning electron microscope. Electropolishing of 1Cr-1Mo-1/4 V steel was performed using Jacquet's solution (acetic anhydride: perchloric acid = 4:1) at a constant current of 0.7 A for 50 s on an area of 50 mm². To ensure that the area density and length distribution of small cracks per unit area do not depend on the size of the measurement area, an area of 5 mm² for the 0.10 N_f and 0.17 N_f SUS304 specimens and 3.75 mm² for each specimen of 0.25 N_f or greater on a longitudinally split section including the specimen center axis was used as the observation area. For the 1Cr-1Mo-1/4 V steel, approximately 2 mm² of the electropolished surface (50 mm²) of each interrupted specimen was selected, including the specimen center axis.

The angle measured here, as shown in Fig. 4.3, was the angle between the straight line connecting the two ends of the crack and perpendicular to the stress axis for a crack of one grain boundary length immediately after initiation, and the angle between the straight line connecting the triple point of the grain boundary and perpendicular to the stress axis for a grain boundary. The crack length and grain boundary length are half-lengths of the projection perpendicular to the stress axis. All cracks in the observation area were measured. The minimum crack half-length measured was 10 μ m in each case, due to the magnification of the optical micrographs and the need to distinguish cracks from polishing flaws and grain boundaries. It was confirmed that the small cracks were uniformly distributed within the specimen.

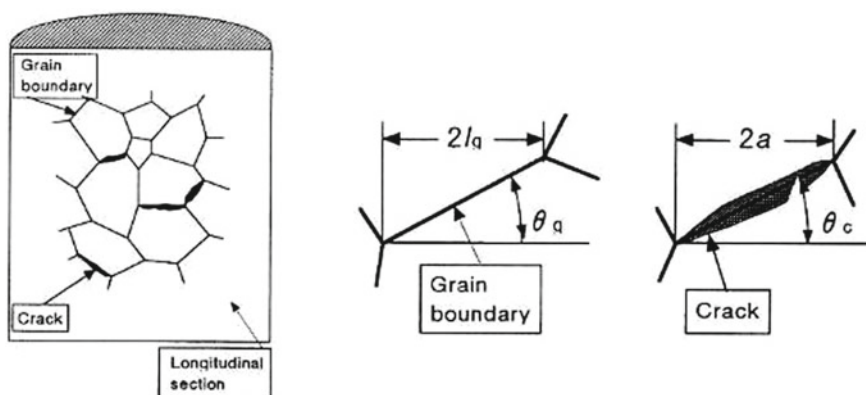


Fig. 4.3 Inclined angles and lengths of crack and grain boundary measured on a longitudinal section of the specimen [8]

4.2.3.2 Three-Dimensional Observation

For the $0.75N_f$ specimen of SUS304 described above, the longitudinally split cross section of the specimen was sequentially polished by buffing and diamond at approximately $3.5\text{ }\mu\text{m}$ depth, and the volume density of small inner cracks was measured by observing the cross-sectional conditions of 20 surfaces based on photographs taken each time with an optical microscope. The total amount of polishing was approximately $75\text{ }\mu\text{m}$. Therefore, the volume observed in three dimensions is $3.75\text{ mm}^2 \times 0.075\text{ mm}$ ($= 0.281\text{ mm}^3$).

4.3 Results and Discussion

4.3.1 Inner Cracks

Figure 4.4a and b present examples of observations of longitudinally split sections of $0.25N_f$, $0.5N_f$, $0.75N_f$, and N_f specimens of SUS304 and 1Cr-1Mo-1/4 V steel, respectively. In both steels, small inner cracks are randomly initiated and grow on the grain boundaries almost perpendicular to the stress axis. During the early stages, the number of cracks remains small, but as the life progresses, numerous small cracks are initiated and grow or coalesce. However, even in the first half of the life, detailed observation of both steels shows that cavities, as shown in Fig. 4.5, are continuously initiated on the grain boundary perpendicular to the stress axis in large numbers. Figure 4.6 shows an optical micrograph of a chemically etched longitudinally split section of a $0.75N_f$ specimen of SUS304. The figure shows that the cracks were initiated and grew in a single grain boundary unit in the direction almost perpendicular to the stress axis and that there were few cases where two cracks that were more than one grain diameter (approximately $50\text{ }\mu\text{m}$) apart in the stress axis direction coalesced.

Figure 4.7 shows the maximum tensile and compressive stresses for both tests. As shown in Fig. 4.7a, the SUS304 steel exhibits cyclic hardening in the early stage of fatigue life, followed by a steady state and a sudden drop in stress at the latter stage of fatigue life, leading to rupture. On the other hand, as shown in Fig. 4.7b, 1Cr-1Mo-1/4 V steel exhibits cyclic softening in the early stage of fatigue life, followed by a gradual decrease in stress without a clear steady state, leading to rapid stress decrease and rupture at the end of fatigue life. The rapid decrease in stress at the end of life for both steels suggests that small cracks formed in the specimen rapidly coalesce to form a main crack at the end of life, which reduces the cross-sectional area of the specimen and leads to rupture.

Figure 4.8 shows the main cracks that caused failure in the N_f specimens. For both steels, the main crack formed in the center of the specimen and consisted of several small cracks with different openings. In combination with the large number

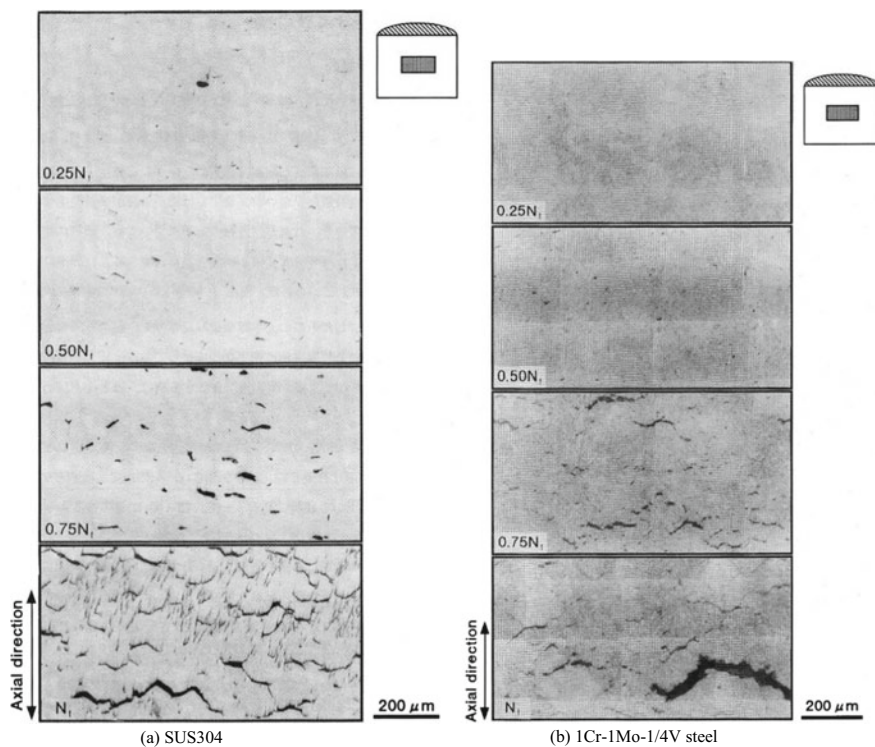


Fig. 4.4 Morphology of cavities and cracks revealed on a longitudinal section of the specimens [8]

of small cracks, it can be expected that the main crack formed from a large number of connected small cracks at the end of the life as described above.

As mentioned above, both steels exhibit typical internal intergranular cracking type fracture. In the following section, the behavior and initiation/growth mechanisms of small inner cracks in Type 304 stainless steel will be discussed in detail.

4.3.2 Change Over Time of Crack Density on Longitudinal Split Cross Section

The areal crack density at the center of the longitudinal section was measured for six interrupted and fractured specimens ranging from $0.1N_f$ to N_f . The areal crack density was characterized as the number of cracks with a half-length of $10\text{ }\mu\text{m}$ or greater per unit area. The variation of the crack density with lifetime is shown in Fig. 4.9 as a black circle ($\bullet$). Cracks start to initiate at the very beginning of the lifetime (approximately $0.17N_f$) and continue to initiate until late in the lifetime. The crack density is small in the initial stage, but quickly increases until the mid-life

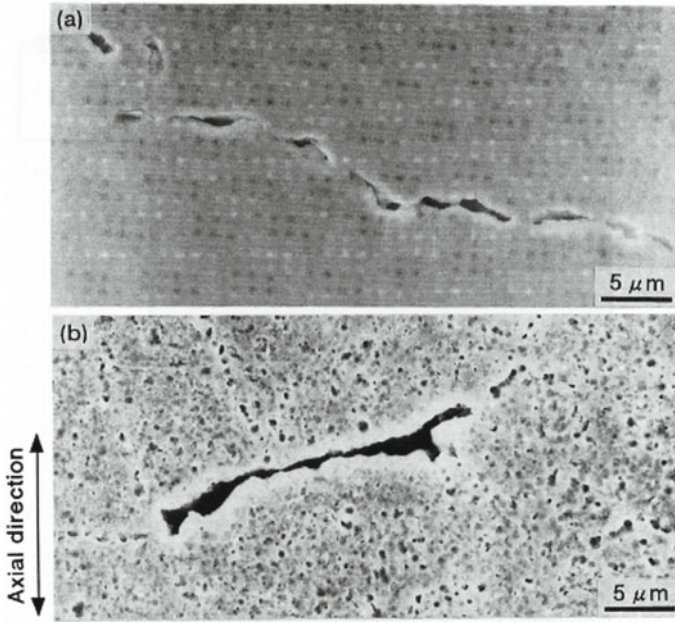
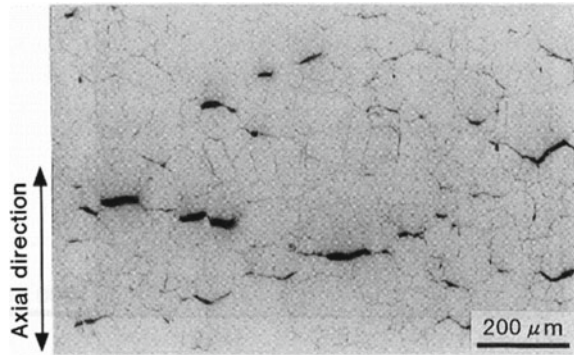


Fig. 4.5 Cavities nucleated along grain boundaries. **a** SuS304. **b** 1Cr-1Mo-1/4 V steel [8]

Fig. 4.6 Grain boundary cracks on a longitudinal section of the fatigued specimen at $0.75N_f$ [8]

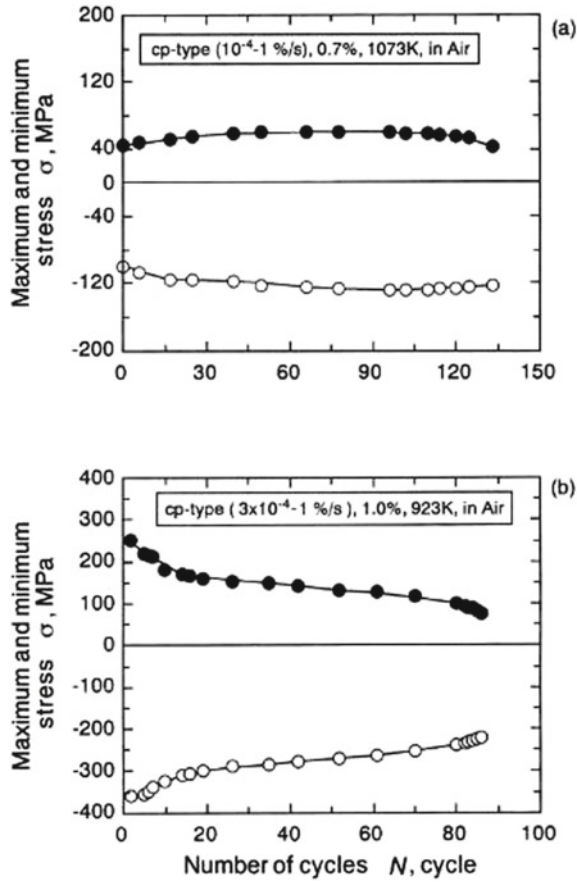


(approximately $0.75N_f$), and then rapidly increases in the later stages, reaching a high density of more than approximately 70 cracks/mm² at failure.

4.3.3 Volume Density of Small Inner Cracks

For the $0.75N_f$ specimen described above, the volume density of small inner cracks (number of cracks per unit volume) was determined based on the method elucidated

Fig. 4.7 Change in peak stresses during fatigue.
a SUS304. **b** 1Cr-1Mo-1/4 V steel [8]



in Sect. 4.2.3.2. As illustrated in Fig. 4.9, the area density of the longitudinal section of the small inner cracks with a half-length of 10 μm or longer is about 35 cracks/ mm^2 , but the volumetric crack density (marked with ■ in the figure) is very high at about 1400 cracks/ mm^2 .

Determining the volume density of small inner cracks, their individual dimensions, and their distribution over time using such a measurement method would require an inordinate amount of effort and would be nearly impossible. Therefore, we will estimate them by numerical simulation based on the crack information on the longitudinal section measured in 4.3.2 (Fig. 4.9) and 4.3.4 (Fig. 4.10) in this chapter. The methodology and results will be detailed in the next chapter.

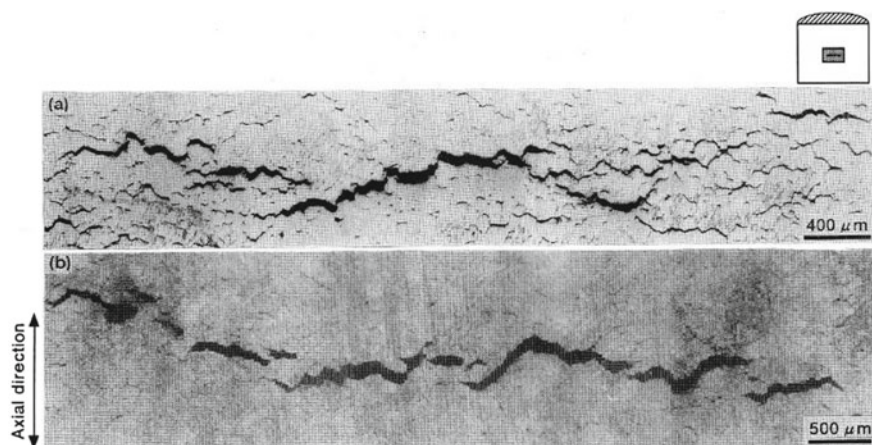


Fig. 4.8 Main cracks generated by coalescence of many aligned small cracks inside the specimens [8]

Fig. 4.9 Change in crack density per unite surface of a longitudinal section. The square mark shows the crack density per unit volume [8]

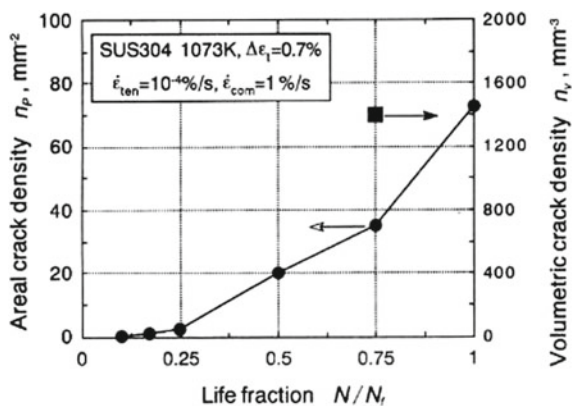
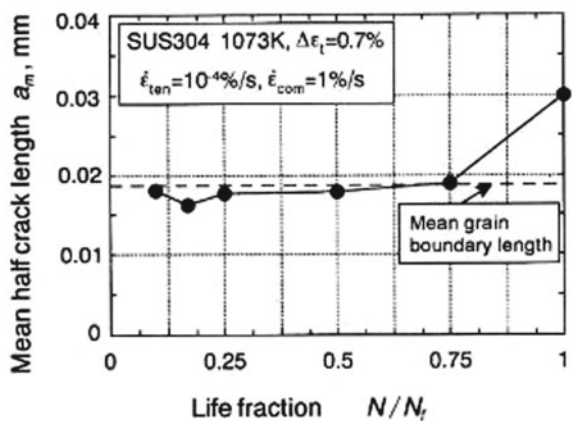


Fig. 4.10 Change in mean length of cracks on a longitudinal section of the interrupted specimens [8]

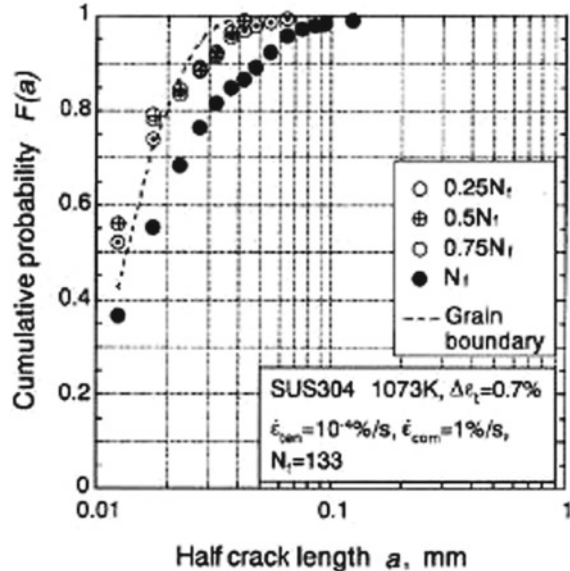


4.3.4 Change in Crack Length Distribution with Time

Figure 4.10 shows the change of the average crack length. The dashed line in the figure represents the average length of the grain boundary. The mean crack length remains consistent up to 75% of the lifetime and is similar to the mean length of the grain boundary, but it increases rapidly in the later stages.

Figure 4.11 shows the change of the crack length distribution in the longitudinally split cross section of the specimen. The crack length distributions of the $0.10N_f$ and $0.17N_f$ specimens were not measured because the number of cracks was small, as mentioned in the previous section. The dashed lines in the figure indicate the distribution of grain boundary lengths of the specimens. However, the measurement of grain boundary lengths corresponds to that of small cracks, and only those with a projected half-length of $10\text{ }\mu\text{m}$ or greater were measured. The figure shows that there is no significant difference in the crack length distribution of the specimens up to 75% of the failure life, which is almost consistent with the grain boundary length distribution. This indicates that the length at which the cavity coalesces to form a crack corresponds to one grain boundary length and is the crack initiation unit that the difficulty of crack initiation is almost independent of the grain boundary length, and that crack initiation is dominant until late in the life with no significant crack growth. The overall crack length distribution shifts toward the longer as life fraction increases, indicating that the cracks are growing.

Fig. 4.11 Distribution of crack length on a longitudinal section and its change during fatigue [8]



4.3.5 Relationship Between Small Inner Cracks and Intergranular Cavities

4.3.5.1 Grain Boundary Cavity Region

As pointed out in Sect. 4.3.1, cavities rarely initiate in isolation, but rather multiple cavities initiate continuously at certain intervals on a grain boundary. Hereafter, the region where such cavities are observed is referred to as the grain boundary cavitation zone. This cavitation zone can be classified into two types: those on grain boundaries that are not adjacent to cracks and those on grain boundaries adjacent to the crack tip.

Figure 4.12a shows SEM images of cavities on grain boundaries which are not adjacent to cracks. Cavities on the order of a few micrometers are rarely found alone, but rather as a series of cavities on the grain boundary, as shown in the photograph. The cavities are rarely quasi-equilibrium [5–7] spheres, but often have a crack-like [5–7] shape. The growth rate of cavities in the grain boundary cavitation zone is considered to be the cause of crack initiation.

Figure 4.13a shows SEM images of cavities on the grain boundary adjacent to the crack tip. This intergranular cavitation zone is the pathway of the crack growth process and shows that the damage near the crack tip is in the process of accumulating in the form of cavity initiation and growth. Therefore, the growth rate of cavities in the intergranular cavitation zone adjacent to the crack is considered to accelerate the growth of the inner crack.

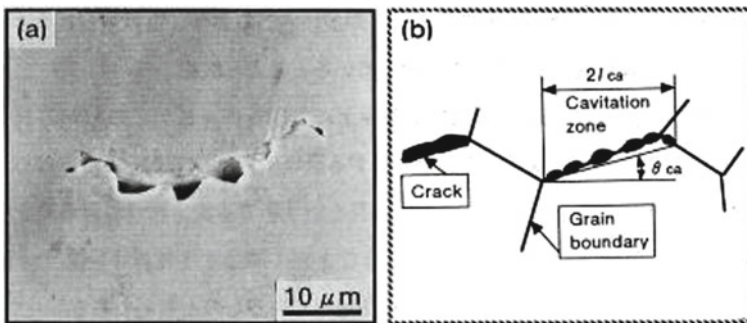


Fig. 4.12 Grain boundary cavities isolated from cracks. **a** SEM Photograph of cavities on a longitudinal section. **b** Schematic illustration of the length and inclined angle of the cavitation zone [8]

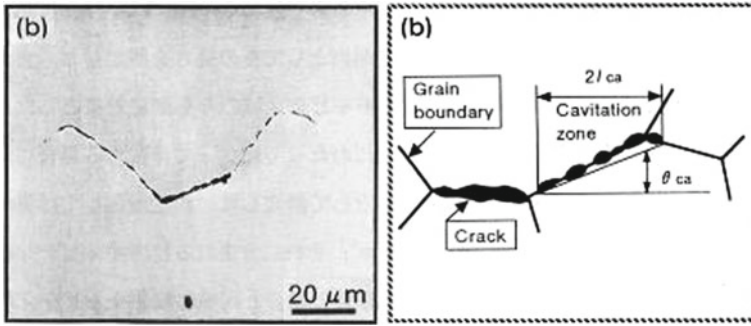


Fig. 4.13 Grain boundary cavities adjacent to a crack tip. **a** SEM Photograph of cavities on a longitudinal section. **b** Schematic illustration of the length and inclined angle of the cavitation zone [8]

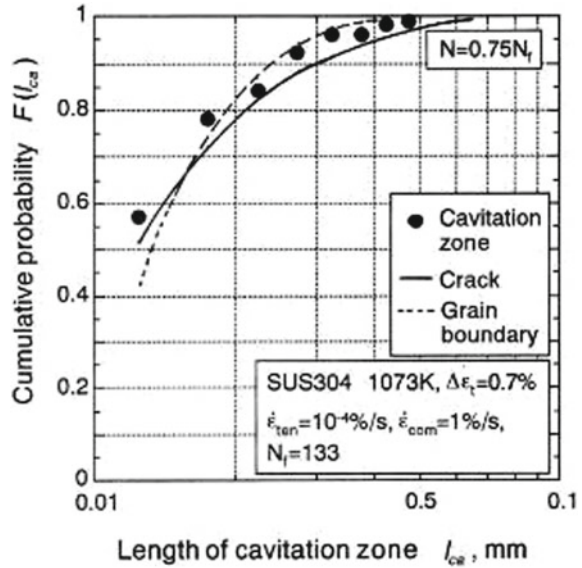
4.3.5.2 Crack Initiation in Intergranular Cavitation Zone

First, the lengths of intergranular cavities that are not adjacent to the crack tip were measured on the $0.75N_f$ specimens to investigate the relationship between the intergranular cavitation zone and crack initiation. Instead of measuring the number of cavities or the length of each cavity, the length of a set of cavities on a grain boundary was measured as a unit, as shown in Fig. 4.12b. As in the crack measurement, only the cavities at the grain boundary with a half-length of 10 μm or longer were measured. However, if a series of cavities existed on more than one grain boundary, the total length of the cavities was measured as the length of a single grain boundary cavity. Five different polished surfaces were selected for the measurement, with a total measurement area of 18.75 mm² and approximately 50 intergranular cavitation zone. The length l_{ca} is half the length of the projection of the straight line connecting the two ends of the intergranular cavity in the direction perpendicular to the stress axis, as in the case of a crack.

Figure 4.14 shows the distribution of the lengths of the intergranular cavities, represented by black circles (•). The solid line in the figure illustrates the distribution of the inner crack length in the same specimen described above, and the dashed line shows the distribution of the grain boundary length. The figure shows that the length distribution of the grain boundary cavitation zone is almost identical to the crack length distribution and the grain boundary length distribution. Consequently, it can be inferred that the grain boundary damage caused by cavities is formed by a unit of grain boundary length. This is the reason why small inner cracks initiate discretely at a unit of grain boundary length.

Figure 4.15a shows the distribution of the angle θ_{ca} between the grain boundary cavitation zone and the direction perpendicular to the stress axis (the angle shown in Fig. 4.12b). Figure 4.15b shows the distribution of the angle θ_c (angle shown in Fig. 4.3) perpendicular to the stress axis for a crack of one grain boundary length immediately after the crack initiation. The angular distributions are almost identical,

Fig. 4.14 Distribution of the length of cavitation zone isolated from cracks observed on a longitudinal section [8]



with the angular distributions of the two cracks being perpendicular to the stress axis (0°). The aforementioned preferential initiation of small inner cracks in the direction perpendicular to the stress axis is considered to be due to the angular dependence of the grain boundary cavitation zone.

4.3.5.3 Crack Growth in Grain Boundary Cavitation Zone

As in the previous section, a $0.75N_f$ specimen was employed to measure the length distribution of the intergranular cavitation zone adjacent to the crack tip, as shown in Fig. 4.13b. The measurement area was 3.75 mm^2 . As in the previous section, as shown in Fig. 4.13b, only grain boundary cavities with a half-length of $10 \text{ }\mu\text{m}$ or longer projected perpendicular to the stress axis were included in the measurement. However, as in the previous section, when the cavitation zone is connected to more than one grain boundary, the length of a grain boundary cavitation zone is defined as the projected length of a straight line connecting the two ends of the cavitation zone. Figure 4.16 shows the length distribution of the intergranular cavitation zone. The dashed line in the figure is the length distribution of the grain boundary. It can be seen that the length distributions of the grain boundary and the intergranular cavity are not much different, although they are slightly longer than the length of the grain boundary. If we consider the formation of the grain boundary cavitation zone adjacent to the crack as a crack growth process, the crack growth is expected to progress on the order of the grain boundary length. In other words, the growth process of small inner cracks, like that of crack initiation, is expected to exhibit a discrete growth behavior on the order of one grain boundary length.

Fig. 4.15 Distributions of the inclined angle of cavitation zone isolated from cracks and the inclined angle of initiated crack on a longitudinal section.
a Cavitation zone, **b** initiated crack [8]

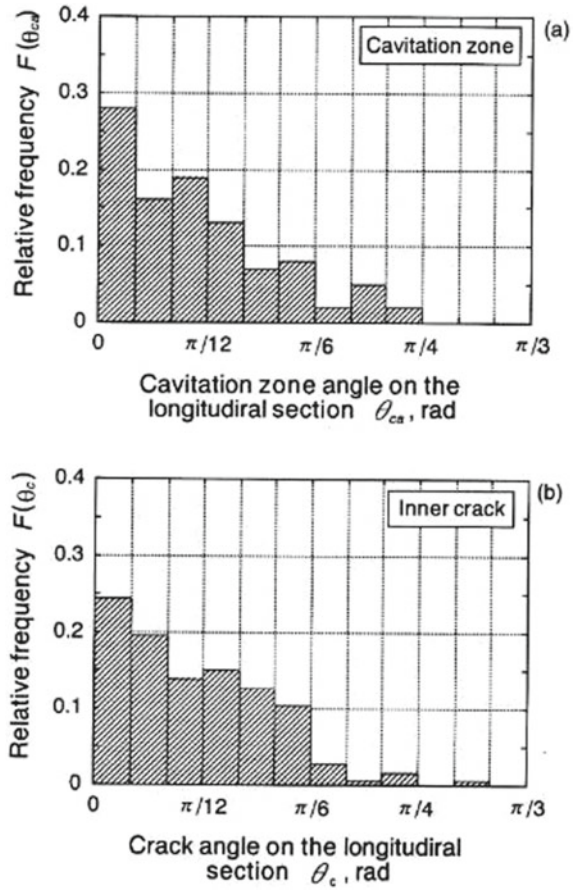


Figure 4.17 shows the distribution of the angle θ_{ca} between the grain boundary cavitation zone adjacent to the crack tip and the direction perpendicular to the stress axis (angle shown in Fig. 4.13b). As with the angle distribution in the grain boundary cavitation zone not adjacent to the crack, the angular distribution of θ_{ca} tends toward the direction perpendicular to the stress axis (0°). Remarkably, both angular distributions exhibit considerable congruence. This angular dependence of the intergranular cavitation zone may be responsible for the preferential growth of small inner cracks in the direction perpendicular to the stress axis, as described above.

The above results suggest that the propagation of small inner cracks is basically similar to small crack initiation, as it occurs through a sequence of initiation, growth, and coalescence of cavities on the grain boundary adjacent to the crack tip [8].

Fig. 4.16 Distribution of the length of cavitation zone adjacent to a crack tip observed on a longitudinal section [8]

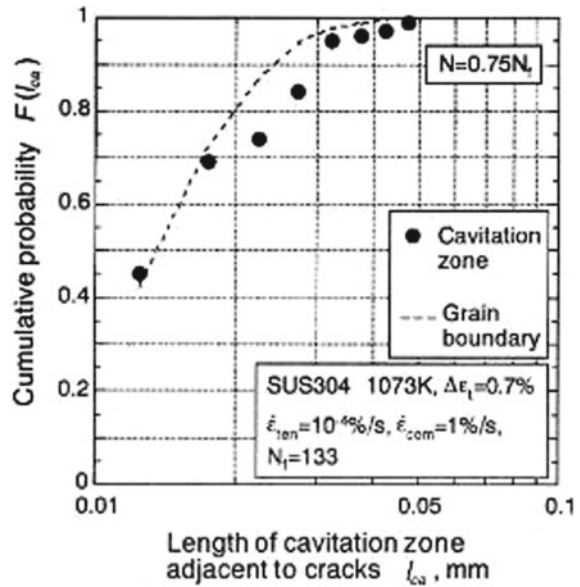
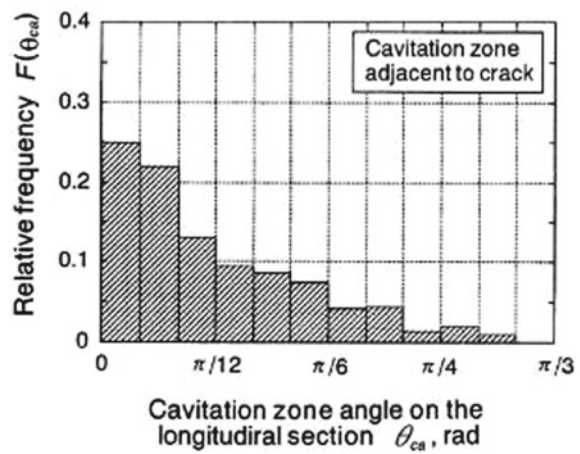


Fig. 4.17 Distribution of the inclined angle of cavitation zone adjacent to a crack on a longitudinal section [8]



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Chapter 5

Basics of Model of Random Fracture Resistance of Grain Boundaries



Abstract In this chapter elucidates modeling and simulation methods of one-dimensional and three-dimensional models of random fracture resistance of grain boundaries for an initiation and propagation of multiple small cracks in creep-fatigue. In the one-dimensional model of random fracture resistance of grain boundaries, the concept of the model is explained, and it is shown that the one-dimensional model can be used to simulate the behavior of small cracks from the early to mid-life. The three-dimensional model incorporates both the microstructural sensitivity of small cracks and the interaction of initiation and propagation of multiple cracks, and thus, enables the simulation up to the later stage of the life, which cannot be conducted by the one-dimensional model. The crack initiation and propagation under a slow-tensile fast-compressive creep-fatigue at 923 K (650 °C) in a vacuum were simulated using the heat-resistant steel with the grain boundaries which were created based on a nucleation-growth model and subsequently whose fracture resistance was set by the model of random fracture resistance. Consequently, it became evident that the experimental results on the crack density, angular distribution of the initiated crack, crack length distribution, and crack growth rate were well simulated.

5.1 Introduction

It is known that creep-fatigue small cracks on the surface of a smooth specimen of SUS 304 exhibit irregular propagation behavior under the influence of microstructure and, in the case that a lot of cracks exist, the cracks grow with coalescence from the middle to the end of the life [1–3]. To clarify the stochastic behavior of small crack initiation and growth due to microstructural effects, the authors performed Monte Carlo simulations using a one-dimensional model of random fracture resistance of grain boundaries. In the model, grain boundaries projected perpendicular to the stress axis are aligned in a straight line and given the fracture resistance and driving force [4]. As a result, the one-dimensional simulation can simulate well the small crack initiation and growth behavior in the early lifetime, but it is difficult to simulate the growth behavior of small cracks in the mid-life time or later when cracks often

interfere and coalesce with each other. It is clear that multidimensional models and simulations are necessary to simulate these behaviors more accurately.

In this chapter, the concept is first explained in terms of a one-dimensional model of random fracture resistance of grain boundaries. Then, the one-dimensional model and the simulations are extended to three dimensions to simulate creep-fatigue small crack initiation and growth behavior up to late in life, incorporating both microstructural effects and interference and coalescence between cracks.

5.2 Results as the Basis of the Model

5.2.1 Conventional Test Results

Observation and measurement of numerous creep-fatigue small cracks on the surface of a smooth specimen of austenitic stainless steel SUS304 revealed the following experimental facts[1, 2].

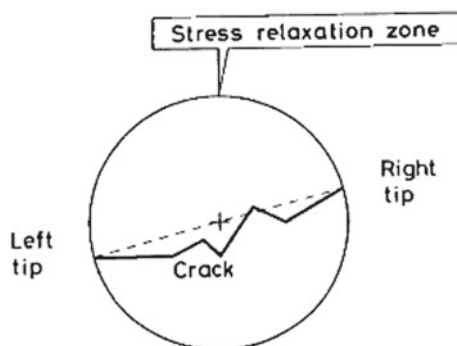
- (1) Creep-cavity initiation, growth, and coalescence cause uniform opening of the grain boundaries between adjacent triple points, resulting in cracks of approximately one grain boundary facet unit.
- (2) All cracks initiate preferentially on the specimen surface, and no inner cracks are observed until late in the life of the specimen.
- (3) The most frequent inclination of the cracks is perpendicular to the stress axis.
- (4) The grain boundaries adjacent to the cracks tend to fracture, and the fracture of the grain boundary becomes the propagation of the crack. That is, the crack propagation is a discrete process consisting of continuous crack initiation.
- (5) Most of the initiated cracks propagate perpendicular to the stress axis. There are few examples of crack propagating in a direction significantly different from perpendicular direction of the stress axis and merging with other cracks due to physical interference.
- (6) The crack density (the number of cracks per unit area) increases almost in proportion to the number of cycles from the beginning to the middle of creep-fatigue.

In addition, detailed measurements of the grain boundaries and the distribution of the length and angle of the cracks in SUS304 specimens are reported [5].

5.2.2 Mechanical Interference Between Multiple Surface Cracks

The stress relaxation in the upper and lower regions near the crack surface is considered to suppress the initiation and growth of new cracks. Therefore, a circular region

Fig. 5.1 Stress relaxation zone [6]



with the diameter at the left–right tip of the crack shown in Fig. 5.1 was set up for the small cracks in c-p type creep-fatigue (vacuum, 923 K, strain range of 1.0%, failure life of 388 cycles), which has a relatively high crack density among creep-fatigue. The difference in the small crack initiation and growth behavior between the inside and outside of the circular region was investigated. As described below, crack initiation and growth are significantly suppressed in this region, which is referred to as the stress relaxation zone.

As a result, all of the small cracks were located outside of the stress relaxation zone of the previously initiated cracks, and no crack initiation was observed inside the stress relaxation zone.

Next, the distributions of crack propagation rate were examined for cracks that were not in the stress relaxation zone of any other crack and cracks that were in the stress relaxation zone of other cracks. Figure 5.2a shows the crack propagation rate distributions for all cracks up to 175 cycles. Until 175 cycles, there were no cracks in other stress relaxation zones, and the cracks were sufficiently far from each other compared to their crack lengths. Figure 5.2b and c show the crack propagation rate distributions for cracks outside and inside the stress relaxation zone, respectively, for more than 175 cycles. Figure 5.2 shows that the cracks inside the stress relaxation zone of other cracks are mostly non-propagated and propagate at very low rate.

5.3 Model of Random Fracture Resistance of Grain Boundaries (One-Dimensional Model) [4]

5.3.1 Modeling

Based on the experimental results in Sect. 5.2, the following modeling is performed.

- (1) All grain boundaries (grain boundaries between adjacent triple points) are projected in the direction perpendicular to the stress axis and arranged in one straight line as shown in Fig. 5.3. However, it is essential to clarify that Fig. 5.3

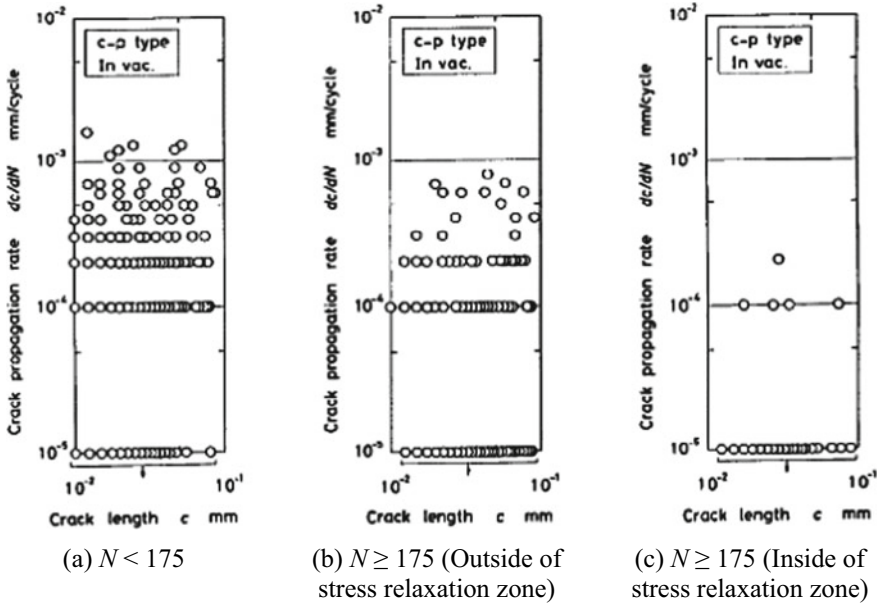
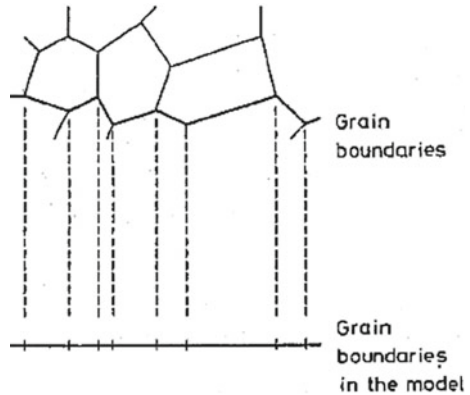


Fig. 5.2 Distribution of crack propagation rate [6]

serves primarily as an illustrative representation for ease of comprehension. The model, in practice, does not employ a direct projection of the actual grain boundaries in this fashion. Instead, it relies on the methodology outlined in Sect. 5.3.2, specifically detailed in (1) and (2).

- (2) Each grain boundary has a value R , which means fracture resistance. This value differs for each grain boundary and gives values with a certain distribution for all grain boundaries.

Fig. 5.3 Grain boundaries arranged in a straight line [4]



- (3) A fracture driving force F_0 is applied to each grain boundary at each cycle unit, and since F_0 damages all grain boundaries, the fracture resistance R of all grain boundaries is decreased by F_0 . It is common to think that the value of F_0 , like R , varies depending on the grain boundary, but in this model, only R has a distribution, and F_0 is a constant value determined by the creep-fatigue conditions.
- (4) When the fracture resistance R of a grain boundary becomes zero or negative, the grain boundary fails, and a crack is initiated.
- (5) The fracture driving force of the grain boundary adjacent to the cracked grain boundary is considered to be a function of the crack length $2c$, $F(2c)$, because the driving force increases with the strength of the stress field caused by the crack. The fracture driving force at grain boundaries not adjacent to the crack is $F(0) = F_0$ as described in (3), and although it is common to consider the value of $F(2c)$ to be different for each grain boundary, in this model it is a function of the crack length $2c$ only.
- (6) When the value of $R - F(2c)$ becomes zero or negative, the grain boundary adjacent to the crack fails and the crack propagates.

5.3.2 Simulation Method

The model in Sect. 5.3.1 is used to simulate the following method. In this study, as an example, a small crack in creep-fatigue caused by slow-fast ($10^{-3} \sim 1\%/s$) triangular strain waveform ($\Delta\epsilon_t = 1\%$) in SUS304, 650 °C, and vacuum is taken.

- (1) The average grain size of the specimen is about 50 μm . The grain size is approximated by a regular hexagon, and the grain boundaries on the modeled straight line are normally distributed with a mean of 25 μm and a standard deviation of 20 μm .
- (2) When the grains are approximated by a regular hexagon with the average grain size, the total number of grain boundaries in the measurement area is (number of grain boundaries) = $2 \times (\text{measurement area})/(\text{hexagonal grain area})$, since one grain contributes to two grain boundaries. The area of the simulated region is the area of the region (1.2 mm $\times$ 1.2 mm) actually used for crack measurement. In this case, the number of grain boundaries modeled is 2216.
- (3) Since the fracture resistance R of a grain boundary is not uniquely determined, it is given as a uniform random number from 1 to 2216 for 2216 grain boundaries.
- (4) Based on the experimental results, the functional form of the fracture driving force $F(2c)$ is given for each grain boundary by the following equation

$$F(2c) = F_0 + K \cdot 2c. \quad (5.1)$$

Note that F_0 and K in the above equations are constants determined by the creep-fatigue conditions. $F(0) = F_0$ is the fracture driving force at grain boundaries with no adjacent cracks.

Figures 5.4 and 5.5 show the effects of the constants F_0 and K on the simulation results when $F(2c)$ is set as in the above equations. Figure 5.4 shows that F_0 is the quantity that controls the crack density change, or crack initiation. Figure 5.5 shows that K is the quantity that controls the crack length distribution, or crack propagation rate. Therefore, the values of F_0 and K can be set by measuring the change in crack density and crack length distribution during the test.

Figure 5.6 illustrates the process of simulation in order to explain the method of simulation described above in a simple manner. In order to avoid skewed uniform

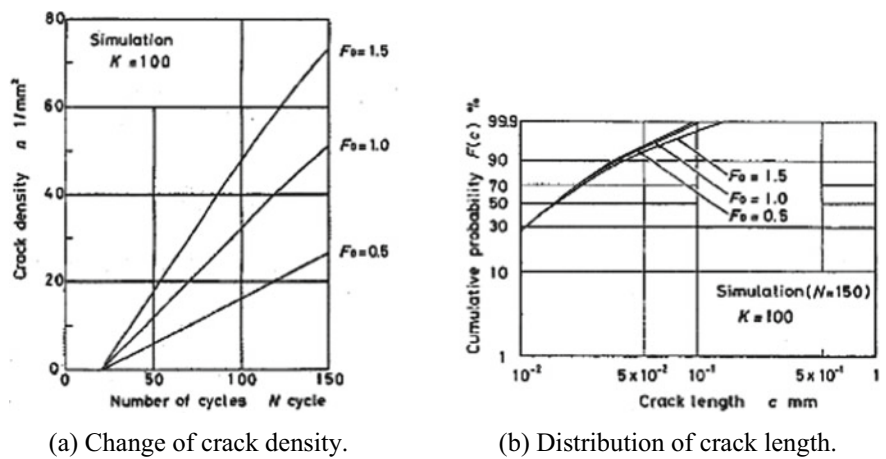


Fig. 5.4 Simulation results (K : constant) [4]

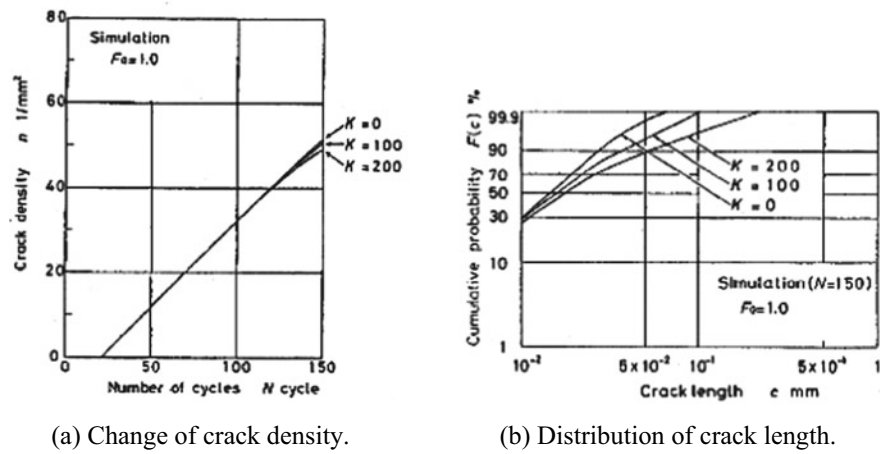


Fig. 5.5 Simulation results (F_0 : constant) [4]

5.4.1.1 Grain Boundary

In order to perform a three-dimensional simulation, it is necessary to create a grain boundary facet that is distributed three-dimensionally including the interior. In this chapter, to avoid the complexity of the model and simulation, only those grain boundary facets that are exposed on the surface (surface grain boundary facets) are created, and the simulation of small crack initiation and propagation is performed only on these surface grain boundary facets.

- (1) Grain boundaries on the surface by grain boundary creation simulation based on the nucleation-growth model [5] are created.
- (2) A three-dimensional surface grain boundary facet by giving all the grain boundaries on the surface an angle θ_g' in the depth direction with a distribution value expressed by the following equation is created.

$$P(\theta_g') = \sin \theta_g' \quad (5.2)$$

The angular distribution expressed in Eq. (5.2) is based on the measurement results [5] of the grain boundary of SUS304 actually used in the creep-fatigue test.

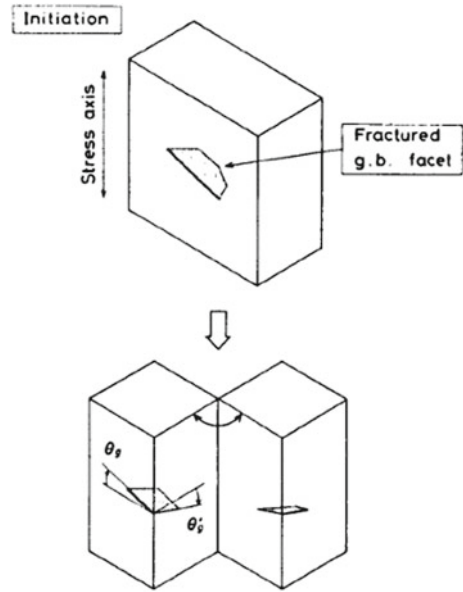
- (3) Each surface grain boundary facet has a value R which means fracture resistance. This value is different for each facet and is given as a value with a certain distribution for all surface grain boundary facets.

5.4.1.2 Crack Initiation

- (1) A fracture driving force F is applied to each surface grain boundary facet for each cycle unit. F decreases the fracture resistance R of all surface grain boundary facets by F because F damages all surface grain boundary facets. The value of F is different for each facet and is a function of the angle $F(\theta_g, \theta_g')$ with respect to the stress axis of each facet. Here, θ_g and θ_g' are the angle at the surface of the facet and the angle in the depth direction, as shown in Fig. 5.7.
- (2) When the fracture resistance R of a surface grain boundary facet becomes zero, the facet is assumed to fracture, and a surface crack is initiated.

Through the utilization of the fracture resistance R and the fracture driving force F as described above, among the randomness of small crack initiation and growth, i.e., the fracture of the grain boundary facets, the variation of strength against cavity or grain boundary slip in each grain boundary facet is mainly represented by R , and the variation of mechanical quantities (stress, strain, etc.) due to the different orientation of each grain boundary facet with respect to the stress axis is mainly represented by F .

Fig. 5.7 Schematic illustration of crack initiation [6]



5.4.1.3 Crack Propagation

- (1) The fracture driving force F for crack initiation is expressed as a function of the projected length c of the surface crack, $K(c)$, this formulation is necessitated by the fact that the fracture driving force for the facet adjacent to the cracked surface grain boundary facet and the facet on the surface that is most closely aligned with the vertical angle to the stress axis, is influenced by the strength of the crack-induced stress field. If the total fracture driving force is D , D is a function of the angle of the surface grain boundary facets to the stress axis θ_g , θ_g' and the surface crack length c and is expressed as below.

$$D(\theta_g, \theta_g', c) = F(\theta_g, \theta_g') + K(c) \quad (5.3)$$

However, the fracture driving force for surface grain boundary facets that are not adjacent to the crack and those that are adjacent but not selected in the propagation direction due to their large angle to the stress axis is $D(0) = F(\theta_g, \theta_g')$. Therefore, $K(c)$ is a function such that it becomes zero when $c = 0$.

- (2) When the value of $R - D(\theta_g, \theta_g', c) \cdot N$ (N : number of cycles) becomes zero, the grain boundary adjacent to the crack fails and the crack propagates.

Similar to the consideration made for F , it is necessary to consider the effect of the angle to the stress axis of the grain boundary facets in the direction of propagation on the value of the fracture driving force K for crack propagation. However, the driving force for surface crack propagation, or in other words, the driving force for fracture

of the grain boundary facets adjacent to the crack, is considered to be averaged by the presence of multiple adjacent facets. That is, a facet with a large angle to the stress axis that is difficult to propagate will be influenced by other facets that are more susceptible to propagation, thereby promoting propagation. On the contrary, facets with higher susceptibility to propagation are restrained by those exhibiting greater resistance to propagation. Considering the average value of the driving force for propagation, the angular dependence of the adjacent grain boundary facets on the value of K is expected to be smaller than that for the fracture driving force F for crack initiation where only one facet is subject to fracture. Therefore, in this model, the fracture driving force F for crack initiation is given as a function of the angle of the facet to the stress axis, and the fracture driving force K for crack propagation is given as a function of only the surface crack length c , $K(c)$.

5.4.1.4 Mutual Interference

Based on the observation of small cracks described in Sect. 5.2, the stress relaxation zone shown in Fig. 5.1 is considered for all initiated cracks, and the following restrictions are given with respect to the fracture driving force.

- (1) Surface grain boundary facets inside the stress relaxation zone are considered to suppress crack initiation by stress relaxation and are not subjected to the fracture driving force $F(\theta_g, \theta_g')$ related to crack initiation.
- (2) Small cracks located inside the stress relaxation zone of a large crack are considered to suppress crack propagation due to the stress relaxation of the large crack, and the surface grain boundary facets adjacent to the small crack are not subjected to the fracture driving force $K(c)$ related to crack propagation.
- (3) Interaction between cracks due to coalescence is not considered.
- (4) When cracks coalesce, they are assumed to form approximately semicircular cracks at that moment.

5.4.2 Simulation Method

The following method is applied to simulate c-p type creep-fatigue small cracks in SUS304 at a temperature of 923 K (650 °C), in a vacuum, and in a strain range of 1%.

- (1) The area of the simulated region is set to 2.0 mm², which is approximately equivalent to the actual crack measurement area of 2.4 mm².
- (2) In the area to be simulated, surface grain boundary facets similar to the grain boundary distribution of the specimen (SUS304) are created using grain boundary creation simulation. The angle in the depth direction of each facet is given by a random number with the distribution function in Eq. (5.2).

The grain boundary length and angle distributions of the created grain boundaries are in good statistical agreement with those of the specimen. For a comparison of the simulated grain boundaries with those of the specimen, please refer to the authors' paper[5].

- (3) The fracture resistance R of the surface grain boundary facets is given as a uniform random number from 1 to 1000 for all surface grain boundary facets.
- (4) The functional forms of the fracture driving forces $F(\theta_g, \theta_g')$ and $K(c)$ are determined as follows.

(i) Fracture driving force for crack initiation

The small crack initiation under the present test conditions is considered to be strongly tensile stress (strain) dependent due to the high frequency of crack initiation in the direction perpendicular to the stress axis. Let the remote axial stress be σ and the perpendicular stress at the grain boundary facet surface be σ_n , where σ_n can be obtained from the following equation.

$$\sigma_n = \sigma / (\tan^2 \theta_g + \tan^2 \theta_g' + 1). \quad (5.4)$$

Since the angular dependence of the crack initiation under the test conditions is expected to be stronger than the angular dependence of the vertical stress, a value m representing the strength of the angular dependence is introduced to give the fracture driving force F related to crack initiation as follows:

$$F(\theta_g, \theta_g') = F_0(\sigma_n/\sigma)^m \quad (5.5)$$

where F_0 and m are constants determined by the creep-fatigue conditions.

(ii) Fracture driving force for crack propagation

In general, the fracture driving force for macroscopic crack propagation in creep is expressed as the creep J-integral J^* . In addition, for creep-fatigue macroscopic crack propagation, the fracture driving force per cycle is expressed by the creep J-integral range ΔJ_c , which is the time integrated value of the tensile half-cycle of the creep J-integral J^* [7]. Therefore, if the specimen is in a homogeneous body with no microstructural effects even for small cracks, the fracture driving force with respect to crack propagation is considered to be represented by J^* and ΔJ_c . Here, the following equations are presented[8, 9] as evaluation equations for the creep J-integral range for small cracks compared to the specimen size.

$$\Delta J_c = \int_0^\tau M_J f(n) \sigma \dot{\epsilon} dt \cdot c \quad (5.6)$$

In the above, τ is the tensile time in one cycle, M_J is the correction factor for crack shape and boundary conditions, and $f(n)$ is a function of creep index n only. Based on this study of the creep J-integral, the fracture driving force $K(c)$ per cycle

for crack propagation in this simulation is given by the following equation, which has a mechanical meaning equivalent to (5.6).

$$K(c) = K_0 c, \quad (5.7)$$

where K_0 is a constant determined by the creep-fatigue conditions.

Now, if the fracture resistance becomes zero in the middle of a cycle, the excess fracture driving force is converted back to the length of the crack after propagation and given to the adjacent grain boundary facets. For example, if a crack propagates for one grain boundary facet in the middle of a cycle, the residual fracture resistance R_2' at the end of the cycle for the second grain boundary facet from the original crack tip, in other words, the grain boundary facet adjacent to the crack tip after propagation, is as follows:

$$R_2' = R_2 - (K_0 - R_1/c) \cdot (c + l_1), \quad (5.8)$$

where R_1 , and l_1 are the fracture resistance and surface length of the grain boundary facets adjacent to the crack before propagation, respectively, and R_2 is the fracture resistance of the grain boundary facets adjacent to the crack after propagation. If a crack propagates for more than two grain boundary facets in one cycle, the above equation is repeated.

5.5 Simulation Results and Discussion

Simulation results of c-p type (in vacuum) creep-fatigue small cracks are presented and compared with experimental results. In the numerical simulations, the following three constants were set.

$$F_0 = 8, K_0 = 90, m = 10. \quad (5.9)$$

The study based on the one-dimensional model [4] shows that F_0 controls the change in crack density from early to mid-life, and K_0 controls the crack length distribution during the same period. The F_0 and K_0 values in Eq. (5.9) were selected so that the crack density and crack length distributions from the early to mid-life are in close agreement with the experimental results. The m value was determined from the angular distribution of the initiated crack.

All simulations were performed five times each, and it was confirmed that there were no significant differences among the simulations. In the actual crack measurements, the minimum crack length was set to 20 μm , which is the length that can be distinguished from grain boundaries and scratches, and therefore, in the simulations, only cracks with a projected length of 20 μm or longer were considered to be cracks.

Figure 5.8 shows the simulation (solid line) and measurement (circled) results for crack density. The simulation results are almost consistent with the measurement

results, not only in the early stage of the lifetime when the F_0 value is determined, but also in the late stage of the lifetime up to 300 cycles. In particular, the saturation and reduction of crack density at the end of life due to the influence of stress relaxation zones, crack coalescence, and other factors are well simulated.

Simulation and measurement results for the distribution of the angle of initiated cracks are shown in Fig. 5.9. It can be seen that the model can simulate the angle dependence of crack initiation by selecting the appropriate m .

Simulation and measurement results for the crack length distribution are shown in Fig. 5.10. The results are in good agreement until late in the lifetime, including the effect of crack coalescence.

The simulation and measurement results concerning the crack propagation rate and crack length are presented in Fig. 5.11. The range of variation in the propagation rate and the fact that the length of the arrested crack is only a few crystal grains in diameter are well represented in the simulations.

Fig. 5.8 Comparison between simulation and experiment in crack density [6]

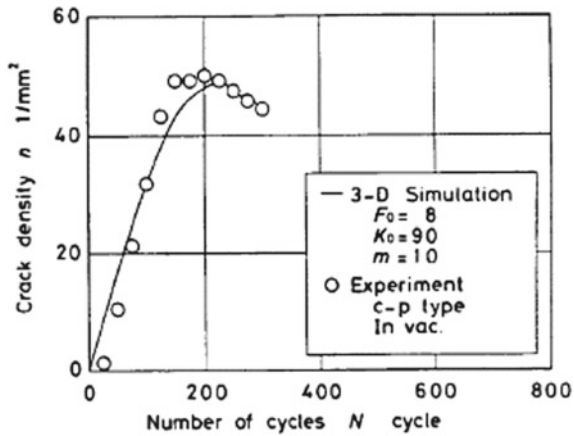
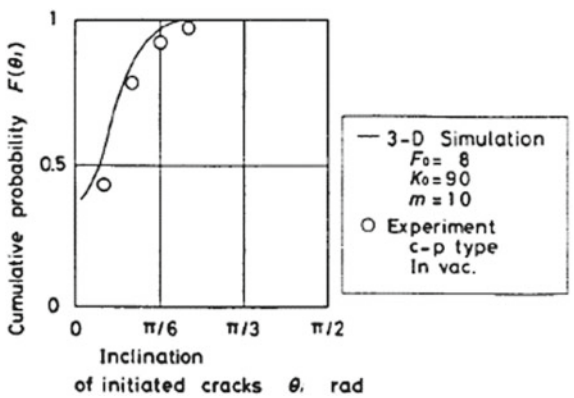


Fig. 5.9 Comparison between simulation and experiment in inclination of initiated cracks [6]



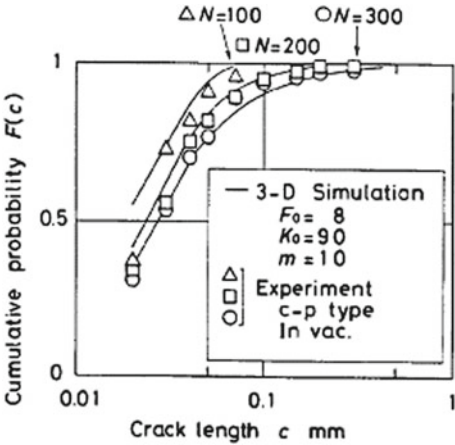


Fig. 5.10 Comparison between simulation and experiment in crack length [6]

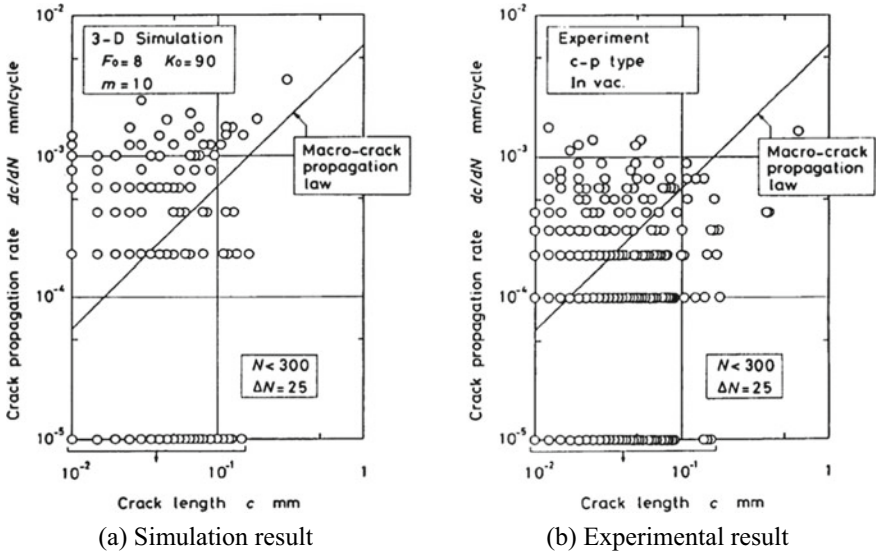


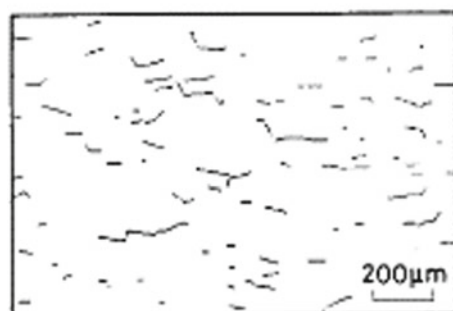
Fig. 5.11 Relationship between crack propagation rate and crack length [6]

Figures 5.12 and 5.13 show simulated and actual surface crack distributions for 100, 200, and 300 cycles, respectively. The distribution patterns of both are visually similar, and this simulation effectively replicates the actual behavior of crack initiation and growth.

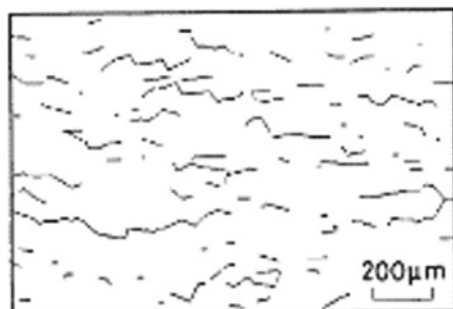
Fig. 5.12 Distribution of small cracks in simulation [6]



(a) $N=100$

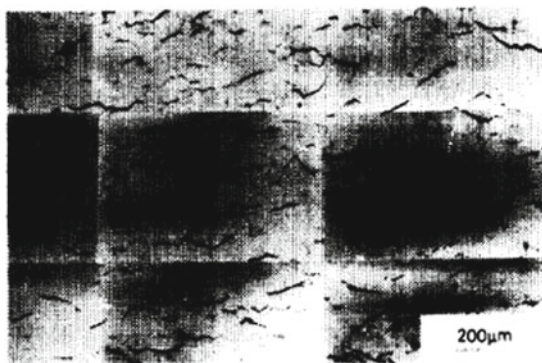


(b) $N=200$

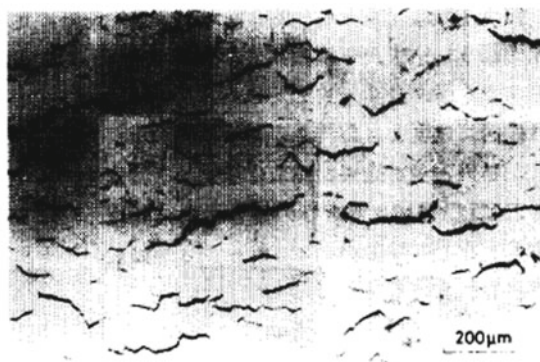


(c) $N=300$

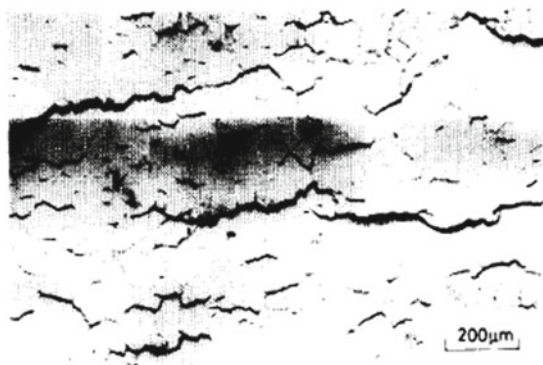
Fig. 5.13 Distribution of small cracks (surfaces of the specimen subjected to c-p type loading in vacuum) [6]



(a) $N=100$



(b) $N=200$



(c) $N=300$

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Chapter 6

Numerical Simulation of the Initiation and Growth of Small Inner Cracks



Abstract In this chapter, regarding small inner cracks in creep-fatigue of austenitic stainless steel SUS304, we proposed a numerical simulation method to estimate the spatial distribution of small inner cracks and its changes that cannot be directly and continuously observed, based on the observations of intergranular cracks on the stress axis longitudinal section of specimens obtained from interrupted tests. The small inner crack model, which is the basis of this simulation, is created by a computer model of a stack of planes (multiplanes) consisting of grain boundary facets projected on a plane perpendicular to the stress axis, and cracks are assumed to initiate and grow on grain boundary facets with a temporal and spatial distribution in units of grain boundary facets. By selecting an appropriate crack initiation driving force F and crack propagation driving force K , the crack density, mean crack length, and crack length distribution actually measured on a longitudinal section can be well reproduced from early to late life.

6.1 Introduction

The internal intergranular cracking type fracture, which is caused by the initiation, growth, and coalescence of small intergranular cracks initiated uniformly inside the material, begins with the initiation and growth of creep cavities dominated by grain boundary diffusion. As in the case of surface intergranular cracking type fracture [1–10], it is important to clarify the process of the crack initiation and the subsequent crack growth. To this end, it is essential to know the temporal and spatial distributions of small cracks. The two-dimensional observation method of cracks on a longitudinal section parallel to the stress axis as carried out in Chap. 3 is the most orthodox and simple method, but it is difficult to accurately grasp the three-dimensionally distributed cracks. In order to overcome this problem, there is a method of accumulating observations while sequentially polishing the cut surfaces, but this requires a lot of effort and time if the cracks and the number of cracks are small.

Given this situation, it is necessary to estimate the spatial three-dimensional distribution of cracks by some new method. In this chapter, a new method for estimating

the distribution and its change with time of many small cracks inside a material using limited information on the longitudinal cross section of a specimen obtained by a creep-fatigue interruption test and numerical simulation. Furthermore, we will use this method to examine the damage change due to small inner cracks during creep-fatigue in SUS304 steel.

6.2 Basic Experimental Facts

In Chap. 4, a high-temperature creep-fatigue test was conducted using a solid round bar smooth specimen of austenitic stainless steel SUS304 (diameter of 10 mm, parallel part length of 20 mm), and small inner cracks on a longitudinal cross section were observed and measured. The following test results were obtained.

- (1) Numerous creep cavities initiate and grow along the grain boundary inside the specimen. Their coalescence results in one grain boundary facet as a crack initiation unit.
- (2) Small cracks initiate randomly in time and space.
- (3) Threshold of initiating small cracks is almost independent of grain boundary length.
- (4) Most cracks initiate and grow preferentially on grain boundary that are perpendicular to the stress axis (i.e., grain boundaries that are perpendicular to the maximum principal stress). In addition, there are almost no cases in which adjacent cracks separated by more than one grain size (approximately 50 μm) in the stress axis direction coalesce.
- (5) The grain boundary adjacent to the initiated crack is more easily fractured than other grain boundary, and the fracture of the grain boundary becomes the propagation of the crack. Similar to crack initiation, crack propagation occurs on a single grain boundary.
- (6) The crack density is uniform inside the specimen, and there is no difference in the crack length distribution or crack angle distribution in any part of the specimen.

6.3 Simulation Method

6.3.1 *Creation of Grain Boundary Facets*

In order to simulate small inner cracks, it is required to obtain a three-dimensional distribution of grain boundary facets. For the purpose of surface crack simulation, a method has been proposed to create two-dimensional grain boundary on the surface by computer [7, 8, 10], and it is also possible to create three-dimensional random grain boundary facets in a similar way. However, this is not a realistic method because

it requires a huge amount of calculation and memory capacity. On the other hand, regarding surface crack simulation, a one-dimensional model [5, 6] in which grain boundary on the surface are distributed linearly and a two-dimensional model [7, 8] of real grain boundary (which is actually three-dimensional model since the angular distribution in the depth direction is taken into account) have been compared. According to the studies, it was concluded that although the differences between the one-dimensional model and two-dimensional model were observed in the latter stage of fatigue life, the behaviors of crack initiation and propagation from early to middle stage of fatigue life can be simulated by the one-dimensional model [10]. Therefore, a similar simplified model can be applied as a first-order approximation to the inner crack simulation in this study. In fact, the observation results for small inner cracks in the previous chapter suggest that it is possible to consider crack initiation and growth on the two-dimensional grain boundary facet projection plane, in which the three-dimensional grain boundary facets are projected onto a plane perpendicular to the stress axis, as shown in Fig. 6.1. Therefore, in this study, the three-dimensionally distributed grain boundary facets are approximated by stacking a large number of grain boundary facet projection planes as shown in Fig. 6.2, and simulations are performed using these grain boundary facet projection planes without considering crack interference between the projection planes.

In order to create the grain boundary facet model shown in Fig. 6.2, it is first necessary to create a grain boundary facet projection plane. This grain boundary facet projection plane was created numerically by computer based on the nucleation and growth model [11, 12] used to create surface grain boundary. The creation method is summarized as follows. A grain boundary facet projection plane (the line segments

Fig. 6.1 Illustration explaining the projection of grain boundary facets on a plane perpendicular to the stress axis [21]

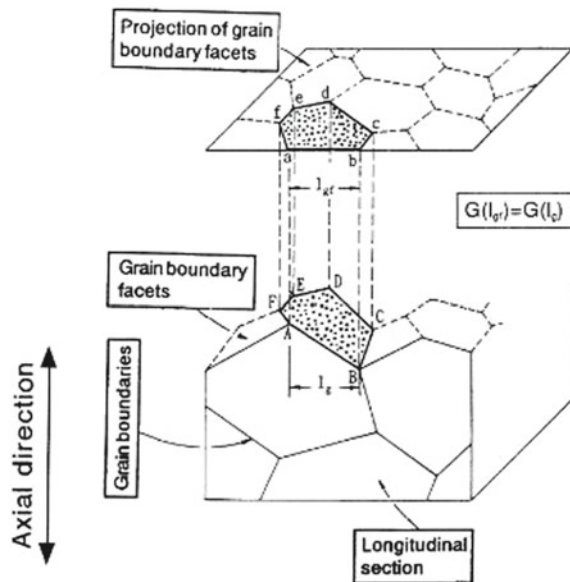
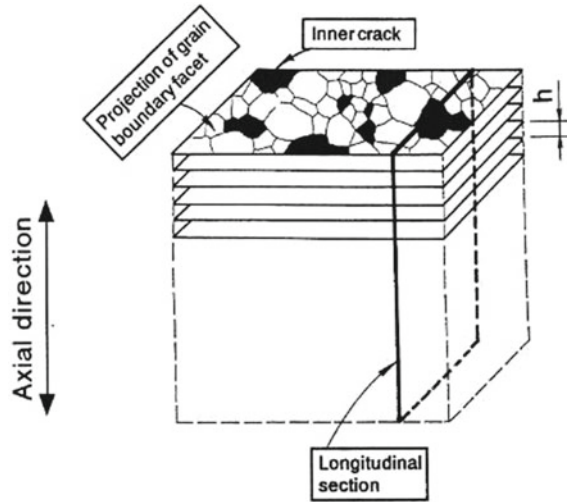


Fig. 6.2 Model of grain boundary facets used for the numerical simulation [21, 22]



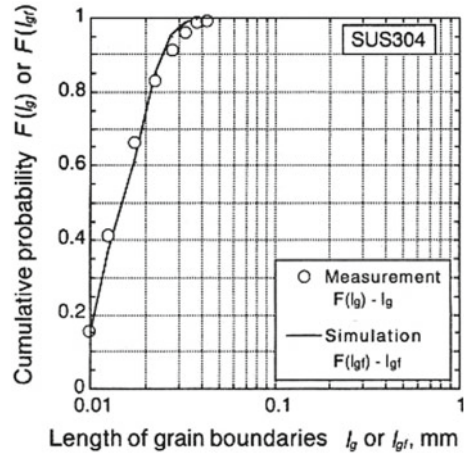
bc, cd, etc., shown in Fig. 6.1 are grain boundary triple lines, not grain boundary) is created here and that is different from the grain and grain boundary in the nucleation-growth model. However, since they are methodologically the same, the names in the nucleation-growth model are marked with a quotation mark (“”) and can be applied as is.

- (1) “Crystal nuclei” (referring to grain boundary facets) are provided in the material using uniform random numbers. Here, the number (density) of “crystal nuclei” per unit area on the projected plane is given by the density of grain boundary facets. In the case of SUS304, the density of grain boundary facets is almost 4 times the density per unit surface area of grains appearing on the longitudinal section.
- (2) Each “crystal nucleus” is given a random “nucleus” generation time P . That is, P is a random variable. Each “crystal nucleus” finishes the nucleation after a time P .
- (3) Each “crystal nucleus” grows isotopically at a speed G after nucleation. Originally, G is a random variable, but since an appropriate grain shape can often be obtained even if G is a constant variable, it can be easily made a constant variable.
- (4) When a “nucleus” encounters another generated phase during its growth, the “crystal nucleus” growth in that direction stops and a “grain boundary” (referring to the triple line of grain boundary facets) is formed.

By using this method and adjusting only P and G , it is possible to obtain projection planes of grain boundary facets with various distributions.

When simulating, as shown in Fig. 6.1, the facet distribution was determined so that the length distribution of the intersection between the cut plane in calculation and the grain boundary facet projection plane $G(l_{gf})$ is equal to the projected length

Fig. 6.3 Distribution of grain boundary facet length in the model and it is compared with the actual grain boundary length [22]



distribution of the grain boundary on the actual specimen in the direction perpendicular to the stress axis $G(l_g)$. At this time, the density of grain boundary facets (projection plane) per unit area was set to 1220 facets/mm², P was a uniform random number from 0 to 1, and G was given as 0.04. Figure 6.3 shows the $G(l_{gt})$ distribution of the created grain boundary facet projection plane and the $G(l_{gt})$ distribution of grain boundary on the actual cut surface of the specimen. The two are in agreement with each other. Figure 6.4a shows the projection plane of the grain boundary facet. Figure 6.4b shows a linear approximation of the triple line of grain boundary facet. In addition, Fig. 6.4c shows the grain boundary facets on the fracture surface of a specimen that was fractured at a low temperature (liquid nitrogen temperature) after interrupting the creep-fatigue test ($0.75N_f$). Figure 6.4d shows the grain boundary facets on the projection plane expressed by straight lines. Figure 6.4b simulates Fig. 6.4d well, and it has been verified in advance that there is no anisotropy in the angle distribution and length distribution of the triplet line on the grain boundary facet projection plane.

In addition, the spacing of the grain boundary facet projection plane h shown in Fig. 6.2 was determined from the following equation.

$$h = (1/n_g)^{1/2} \quad (6.1)$$

Here, n_g is the number of grains per unit area on the cut surface. Substituting the measured grain density of the SUS304 steel ($n_g = 300 \text{ mm}^{-2}$) into (6.1), the spacing h between the grain boundary facet projection planes is approximately 60 μm .

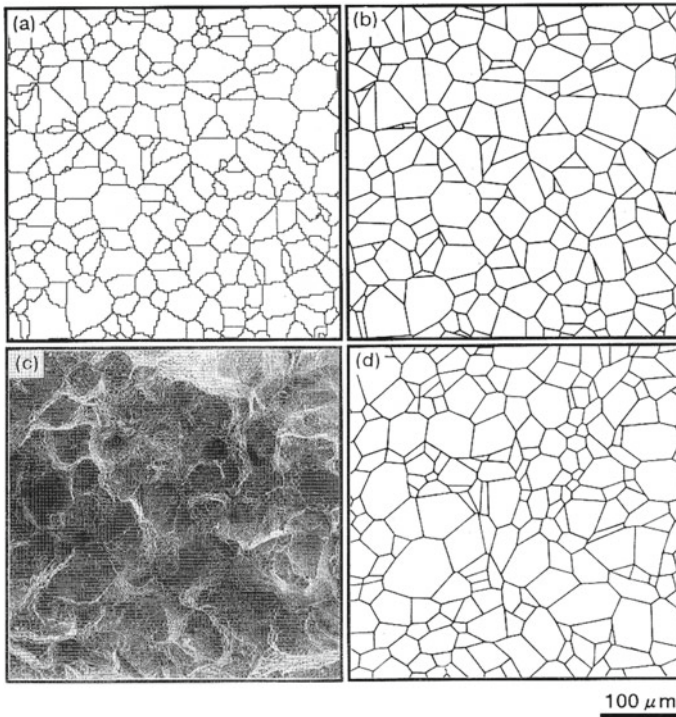


Fig. 6.4 **a** Projection of grain boundary facets obtained by the grain growth simulation. **b** Straight grain boundary facets. **c** Fracture surface. **d** Actual grain boundary projection obtained from **c** [22]

6.3.2 Small Inner Crack Initiation and Growth Model

In general, randomness in the initiation and growth of small cracks is considered to be caused by inhomogeneity of the material (for example, variations in the strength of each grain boundary facet against the initiation and growth of creep cavities), that is, inhomogeneity in the “resistance” to fracture, and by external factors such as the randomness of external forces (for example, variations in mechanical quantities due to differences in the angle of each grain boundary facet to the stress axis and differences in restraint by surrounding grains), that is, the randomness of “driving force” for fracture. The model proposed here is methodologically similar to previously presented models [5–8] regarding surface intergranular cracking type fracture, which is another fracture type of creep-fatigue. The temporal and spatial randomness of crack initiation and growth is analyzed using the Monte Carlo method [13], and its change over time is analyzed using damage mechanics methods [14–17]. Specifically, the details are as follows.

- (1) Fracture resistance value R is given to each grain boundary facet. R is a uniform random number between 0 and 1.

- (2) Fracture driving force F is applied to each grain boundary facet related to crack initiation at each cycle unit. Due to the damage caused by F , the fracture resistance value R of all grain boundary facets decreases by F for each cycle unit.
- (3) When the fracture resistance value R of a certain grain boundary facet becomes zero, that facet is fractured and a small inner crack with the size of one grain boundary facet is initiated.
- (4) The fracture driving force acting on the surrounding grain boundary adjacent to the initiated crack increases due to the stress concentration caused by the crack. Assuming that the fracture driving force is proportional to the equivalent radius r of the crack, the total fracture driving force D is given by the following formula.

$$D = F + K \cdot r \quad (6.2)$$

$$r = (A/\pi)^{1/2} \quad (A : \text{crack area}) \quad (6.3)$$

Here, K is called the fracture driving force related to crack propagation. In general, F and K each depend on the test conditions (e.g., tensile strain rate, compressive strain rate, temperature, etc.) and the angle of the grain boundary facets with respect to the stress axis. However, in this study, F and K are assumed to be constant values for all grain boundary facets, and the irregularity is expressed by the variation in the fracture resistance value R . Now, because of the damage caused by D , the fracture resistance value R of all grain boundary facets adjacent to the crack decreases by D for each cycle unit.

- (5) When the fracture resistance value R of a certain grain boundary facet adjacent to a crack becomes zero, that grain boundary facet is fractured and the crack propagates by one grain boundary facet. However, if the fracture resistance value becomes zero in the middle of one cycle, the excess fracture driving force is converted back to the equivalent crack length after propagation and is applied to the adjacent grain boundary facets.

Figure 6.5 shows a flow chart for simulating the initiation and growth behavior of small inner cracks using a computer according to the above stochastic model.

6.3.3 Simulation Target

Under the same conditions as those for SUS304 in Chap. 4 which exhibited a distinct inner cracking type fracture (in the atmosphere, temperature of 1073 K (800 °C), strain range $\Delta\epsilon_t = 0.7\%$, cp-type ($10^{-4} \sim 1\%/s$) triangular strain waveform test conditions), interrupted specimens with life fraction of 10% (13 cycles), 17% (22 cycles), 25% (33 cycles), 50% (66 cycles), 75% (100 cycles), and 100% (133 cycles) were

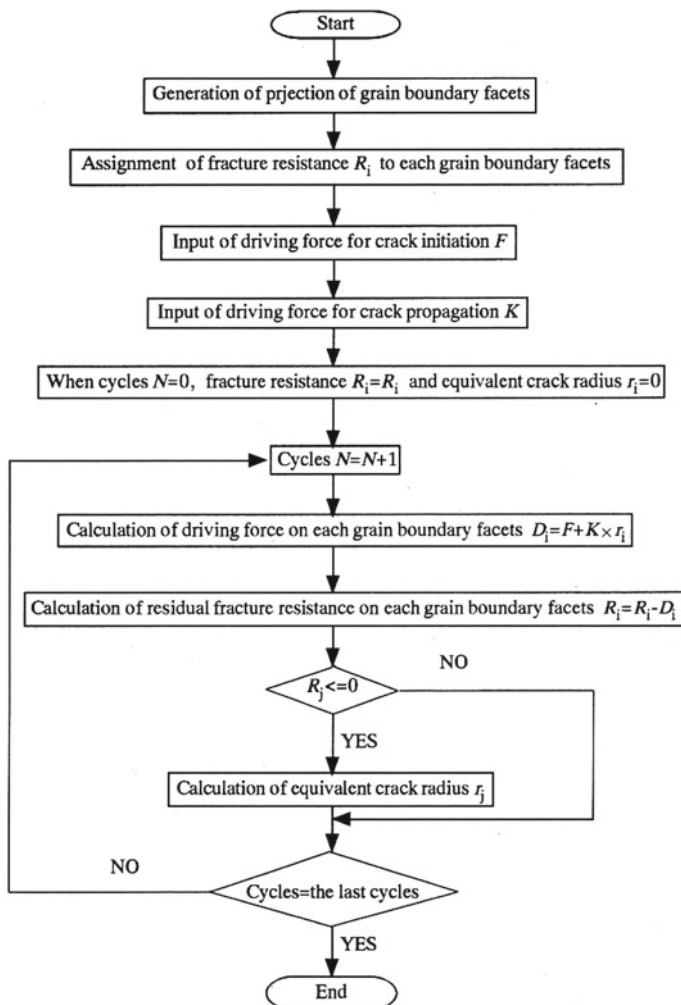


Fig. 6.5 Flowchart of simulation of small inner cracks [22]

prepared. We then observed the changes in the number of inner cracks and the cumulative probability distribution of crack length on these longitudinal cross sections as creep-fatigue progressed. However, the minimum value of the observed crack length (half-length projected on a line perpendicular to the stress axis; hereinafter all crack lengths are expressed as this) was $10 \mu\text{m}$ in all cases.

Figure 6.6a shows the appearance of inner cracks on a longitudinally split cross section. Figures 6.7, 6.8 and 6.9 show the results of experiments regarding the cumulative distribution of crack density per unit area, average crack half-length, and crack length on the longitudinal section of the specimen. The simulation was conducted for this experiment. The experimental results in Fig. 6.9 are the cumulative frequencies

of all cracks (12 cracks when $N = 33$, 48 cracks when $N = 66$, 83 cracks when $N = 100$, and 173 cracks when $N = 133$) for each crack length.

In this study, the volume of the simulated region was selected to be $1 \times 1 \times 1 \text{ mm}^3$, as the volume density per unit volume and small crack radius distribution do not depend on the volume of the calculation region. In addition, although the number of grain boundary facet projection planes in this region is 16, for comparison, we also conducted simulations with four times as many facet projection planes and confirmed that there was no statistically significant difference. In the experimental measurement of inner small cracks, only those whose projected half-length in the direction perpendicular to the stress axis was $10 \text{ }\mu\text{m}$ or more were considered. In order to correspond to this result in the simulation, in the following section, we will

Fig. 6.6 Morphology of inner small cracks revealed on a longitudinal section of the specimens [21, 22]

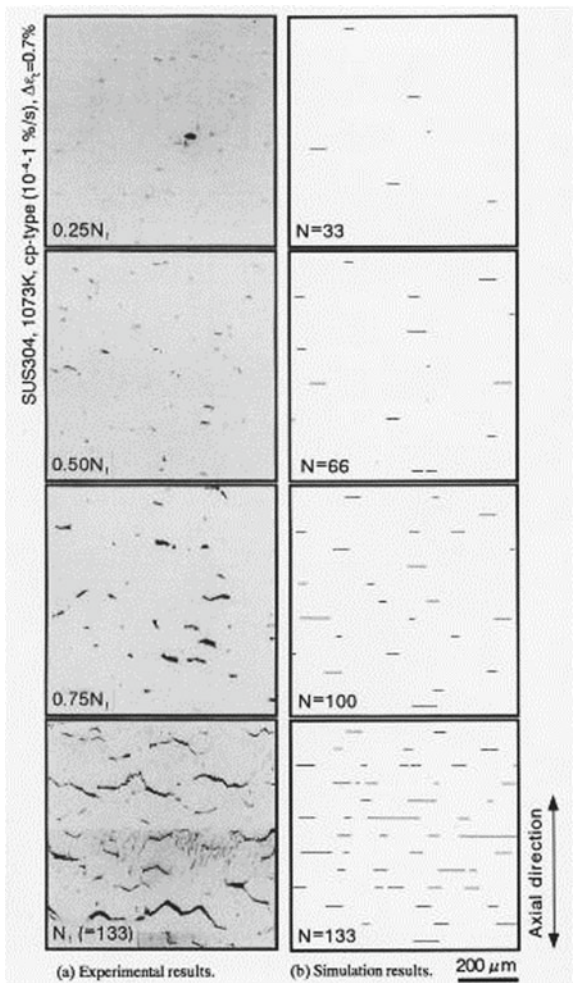


Fig. 6.7 Change in crack density, n_p , observed on a longitudinal section. The line shows the simulation result [21]

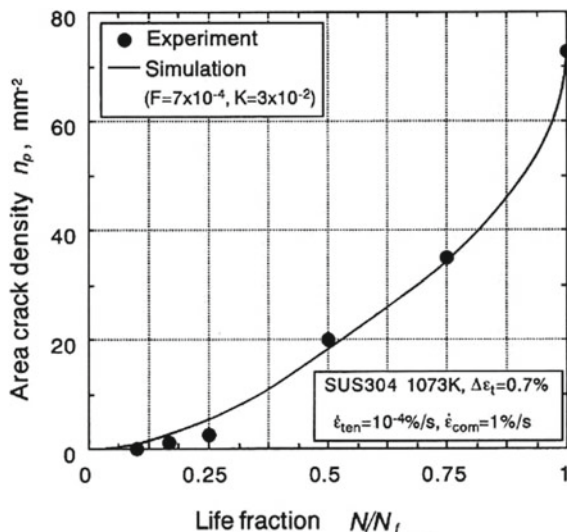
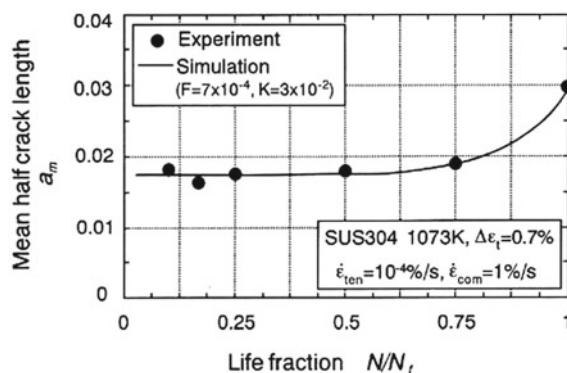


Fig. 6.8 Change in mean crack length, a_m , observed on a longitudinal section. The line shows the simulation result [21]



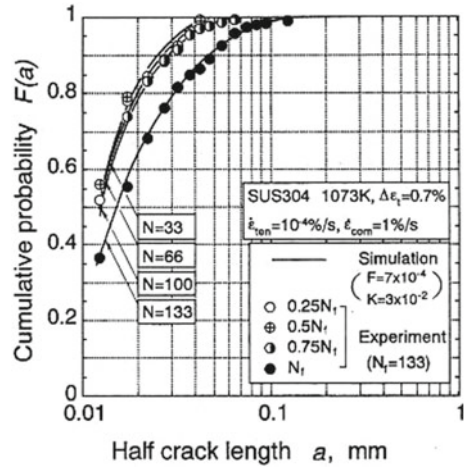
exclude cracks with a half-length of 10 μm or less that appeared on the longitudinal section, and show the results for cracks of 10 μm or more.

6.4 Simulation Results and Discussion

6.4.1 Crack Distribution on Longitudinal Section

Using the simulation method described above, the values of the fracture driving force F for crack initiation and the fracture driving force K for crack propagation were determined from the crack measurement results on the longitudinal sections shown

Fig. 6.9 Distribution of inner crack length observed on a longitudinal section. The lines show the simulation result [21]



in Figs. 6.7 and 6.8. The result is as follows:

$$F = 7 \times 10^{-4}, \quad K = 3 \times 10^{-2} \quad (6.4)$$

In Figs. 6.7, 6.8, and 6.9, the solid lines are the simulation results of the area crack density n_p , the mean half crack length a_m , and the cumulative probability of half crack length $F(a)$ on the longitudinal section of the specimen using the values of F and K in (6.6). Not only changes in crack density and mean crack length from the early life to failure, but also changes in crack length distribution reproduce the test results well. In Fig. 6.6, the simulated crack distribution (b) is in good agreement with the experimental observation results (a) of inner cracks on the longitudinal section. These results show that this simulation can accurately reproduce the behavior of cracks on arbitrary cross sections and suggest that the spatial distribution of cracks can be estimated statistically.

6.4.2 Estimation of Spatial Distribution of Inner Cracks

Figure 6.10 shows an example of the spatial distribution of small inner cracks obtained by simulation ($N = 0.75N_f$, part of the calculation area: $0.4 \times 0.4 \times 1 \text{ mm}^3$). A large number of small inner cracks are uniformly distributed. Figure 6.11 shows the simulation results for the inner crack density per unit volume. In the figure, the results for cracks with a crack radius (equivalent radius according to (6.3)) of $10 \text{ } \mu\text{m}$ or more are shown. The solid symbol in the figure is the experimental results of measuring the three-dimensional density of inner cracks from photographs taken with an optical microscope each time the longitudinal sections of the $0.75N_f$ interrupted specimen under the present test conditions. The cross sections were sequentially buffed and

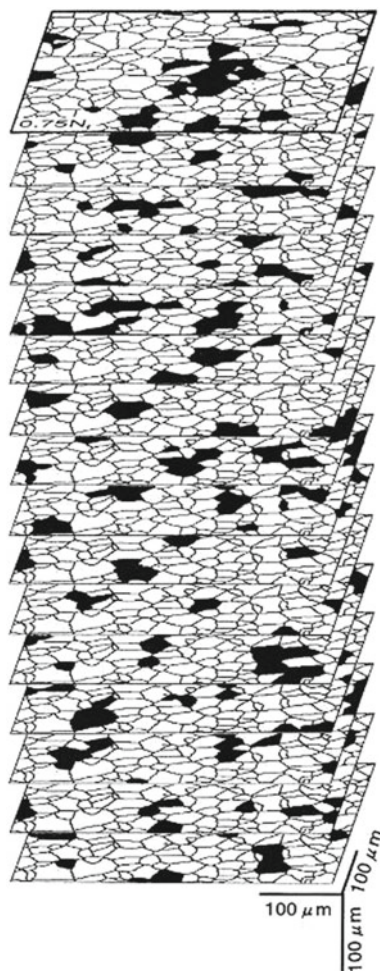
diamond polished at approximately $3.5\text{ }\mu\text{m}$ intervals to a depth of approximately $75\text{ }\mu\text{m}$ as described in Chap. 4, and each time the cross sections were observed by an optical microscope. Because of this measurement method, the volume of the measurement area is limited, and polishing causes errors in determining whether the cracks are the same or not, making it difficult to obtain accurate volume density. Therefore, the solid symbol in Fig. 6.11 can be interpreted as an approximate value of the volumetric crack density. The experimental approximated result and the simulation results are in relatively agreement, demonstrating the validity of estimating the spatial distribution of cracks using this simulation. In addition, until the middle of the life, the volumetric crack density increases almost in proportion to the number of cycles. In the later stages, the area crack density on the longitudinal section shown in Fig. 6.7 increases rapidly, on the other hand, the volumetric crack density is saturated. This is because, as already explained, as the crack radius increases, the probability of a crack being cut by a longitudinal split increases, and the number of cracks in the cross section apparently increases. In this way, the area crack density on a cross section does not necessarily correspond to the spatial volumetric crack density. Furthermore, the area density on the longitudinal section ultimately exceeds 70 cracks/mm^2 , but when converted to volumetric density, it becomes an extremely high density of 1200 cracks/mm^3 .

Figures 6.12 and 6.13 show the mean crack radius r_m , the cumulative probability distribution $F(r)$ of crack radius r estimated by simulation in the inside of the specimen, and their changes over time. Here, the crack radius is the equivalent radius obtained from (6.3). While the crack radius and its distribution hardly change until the middle of the life, the crack radius rapidly increases and the proportion of long cracks increases in the latter half of the life due to the effect of crack growth and coalescence. The probability of a crack appearing in the longitudinal section depends on the crack size and shape, and this has a large effect on the difference between the measurement results on the longitudinal section and the spatial distribution. The appearance of large cracks with irregular shapes due to crack coalescence is considered to have increased the difference between crack distribution on the longitudinal section and the spatial crack distribution at the end of the life.

In continuum damage mechanics [14–17], the creep-fatigue damage state of a material is expressed using the damage internal variable ω . Generally, this damage variable is defined as the rate of reduction in effective area due to the damage such as cavities and small cracks. Figure 6.14 shows the change over time in the rate of decrease in the effective area due to these small inner cracks, and it can be seen that the area ratio of small inner cracks reaches 0.3 at the specimen failure ($N = 133$ cycles). This corresponds to the ratio of the crack area on the grain boundary facet projection plane to the entire projection plane. In other words, it can be assumed that approximately 30% of the cross-sectional area of the specimen is occupied by cracks at the failure. Furthermore, it can be seen that when the crack area ratio exceeds 0.3, the specimen will completely fracture with a short remaining life.

Figure 6.15 shows an example of simulation results regarding crack initiation and growth on the grain boundary facet projection plane. From the figure, the formation process of large cracks can be recognized continuously.

Fig. 6.10 Simulation morphology of the grain boundary facets and small inner cracks (black parts) [21]



6.4.3 *Analysis of Small Inner Crack Initiation and Growth Characteristics by Simulation*

6.4.3.1 **Characteristics of Small Inner Crack Initiation**

As revealed by the above simulations, the initiation of small inner cracks plays a dominant role in the creep-fatigue of the target SUS304, and the crack growth is caused by the coalescence of adjacent initiated cracks. It looks like this. Therefore, we attempted to simulate the case that the unstable fracture is caused only by their initiation and coalescence without any individual growth of the cracks.

The simulation conditions are as follows:

Fig. 6.11 Change in volumetric crack density, n_v , obtained by the simulation [21]

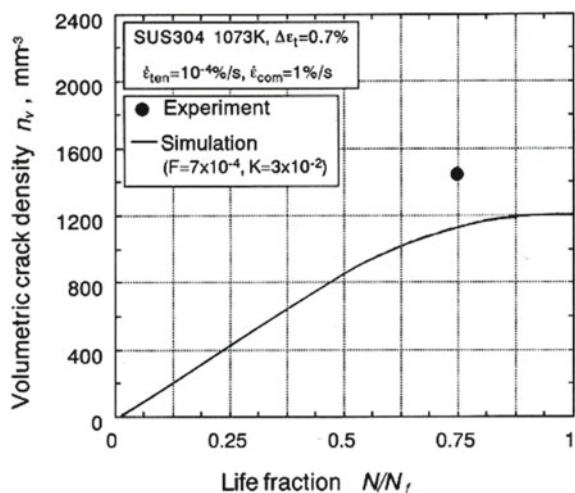


Fig. 6.12 Change in mean crack radius, r_m , obtained by the simulation [21]

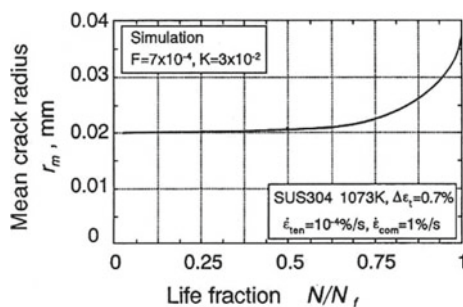


Fig. 6.13 Distribution of inner crack radius obtained by the simulation [22]

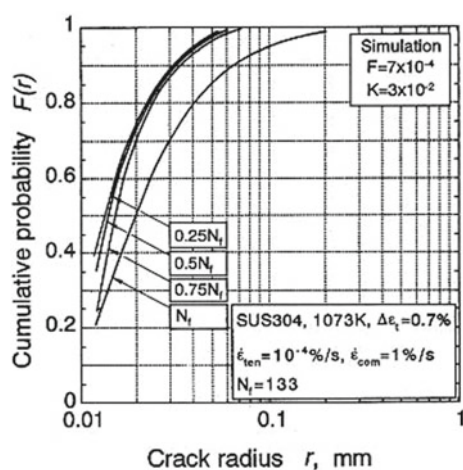


Fig. 6.14 Change in inner crack area fraction, a , obtained by the simulation [22]

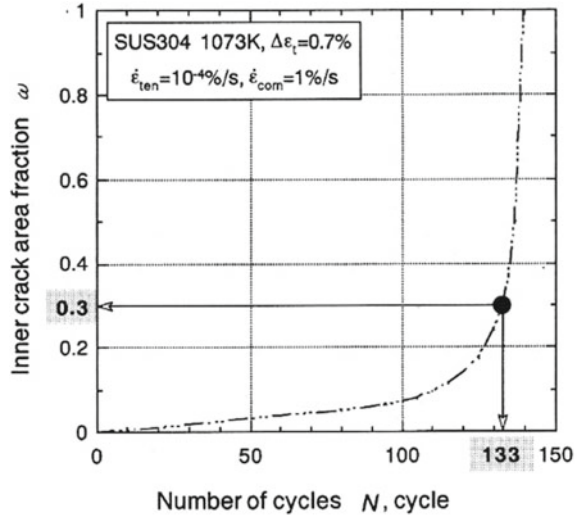
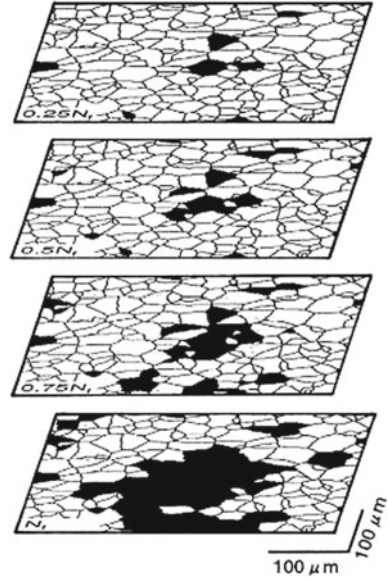


Fig. 6.15 Simulation morphology of the grain boundary facets and small inner cracks (black parts) projected on a plane perpendicular to the stress axis [22]



$$F = 7 \times 10^{-4}, \quad K = 0. \quad (6.5)$$

The crack propagation driving force is replaced with $K = 0$. Figure 6.16 shows the area density and mean half crack length on a longitudinal section based on the simulation, and Fig. 6.17 shows the changes in volumetric density and mean crack radius using dashed lines. The solid lines shown for comparison in the same figure

are the simulation results of Figs. 6.7, 6.8, 6.11, and 6.12 ($K = 3 \times 10^{-2}$). From the beginning to the middle of the life (approximately 70% of the failure life), there is almost no difference between the simulation results under conditions that take into account the above-mentioned crack propagation driving force ($K = 3 \times 10^{-2}$) and those based on (6.5). Both simulation results agree well with the actual measurement results. In other words, this shows that the initiation of small cracks is dominant until the middle of the life and that the propagation driving force is small compared to the initiation driving force. In addition, when the simulation is performed assuming that cracks initiate from the beginning of fatigue, the results are in close agreement with the experimental results, indicating that small internal intergranular cracks in creep-fatigue initiate from an extremely early stage.

From the middle to the end of life, there is a large difference between the results estimated without considering the propagation driving force and those with considering the propagation driving force and actual measurement results. For example, the volumetric crack density estimated without considering the propagation driving force at $N = 133$ cycles (failure life) exceeds 1600 cracks/mm³, which is much higher than the 1200 cracks/mm³ estimated with considering the propagation driving force.

Fig. 6.16 Comparison of simulation results between (1) $F = 7 \times 10^{-4}$, $K = 3 \times 10^{-2}$, and (2) $F = 7 \times 10^{-4}$, $K = 0$, observed on a longitudinal section.
a Change in areal crack density, n_p . **b** Change in mean crack length, a_m [21]

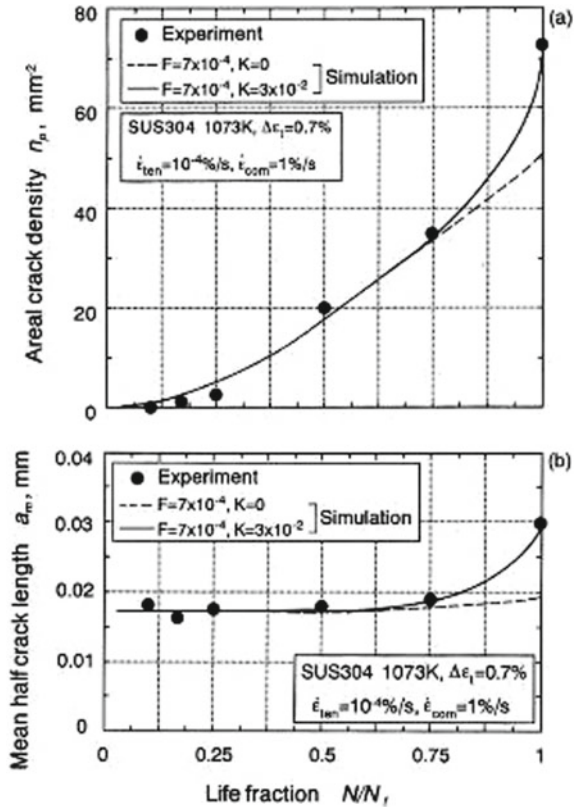
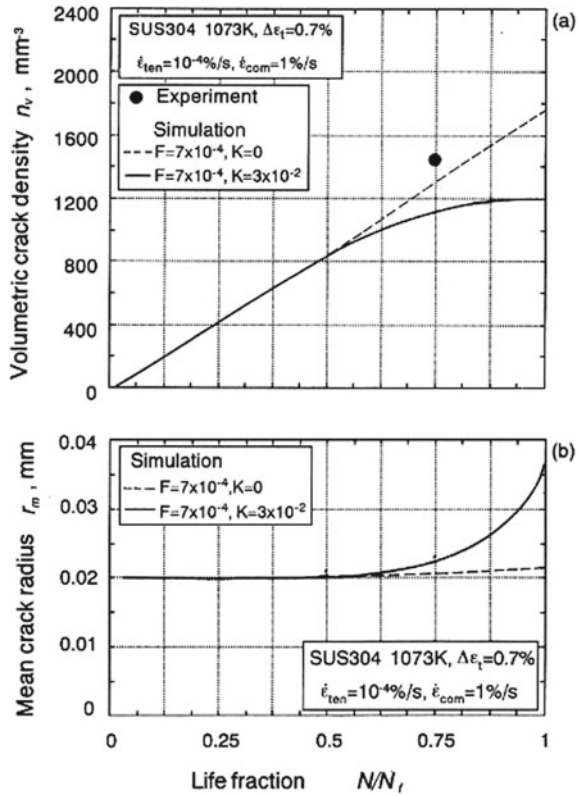


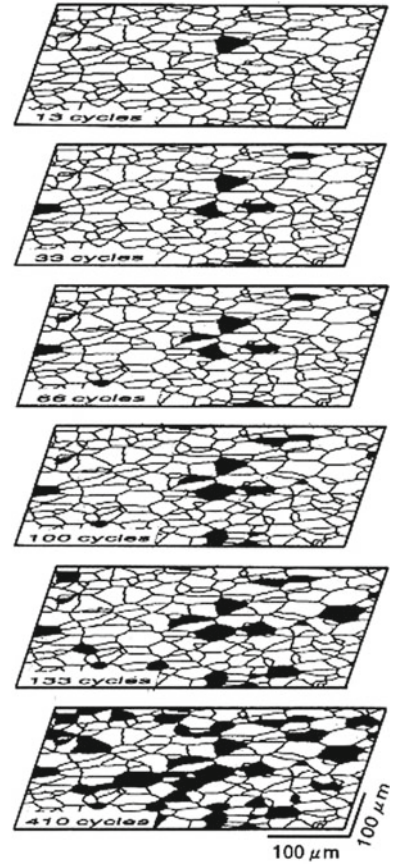
Fig. 6.17 Comparison of simulation results between (1) $F = 7 \times 10^{-4}$, $K = 3 \times 10^{-2}$, and (2) $F = 7 \times 10^{-4}$, $K = 0$. **a** Change in volumetric crack density, n_v , **b** Change in mean crack radius, r_m [21]



This indicates that the crack coalescence with crack growth occurs more frequently than that without crack growth, and as the crack radius increases, the number of cracks, that is, the volumetric crack density, decreases. However, according to this estimation result without considering the crack propagation, the area ratio of small inner cracks at $N = 133$ cycles is only 0.05. To reach the crack area ratio at failure of 0.3 based on the estimation result considering crack propagation mentioned above, it takes approximately 410 cycles (approximately three times the failure life N_f under the test conditions).

Figure 6.18 shows an example of simulation results regarding crack initiation behavior on the grain boundary facet projection plane, as another example of the results estimated without considering crack propagation. Comparing with Fig. 6.15, it is possible to visually recognize the difference in the growth behavior of small inner crack after the middle of the life.

Fig. 6.18 Simulation morphology of the grain boundary and small inner cracks (black parts) project on a plane perpendicular to the stress axis ($F = 7 \times 10^{-4}$, $K = 0$) [22]



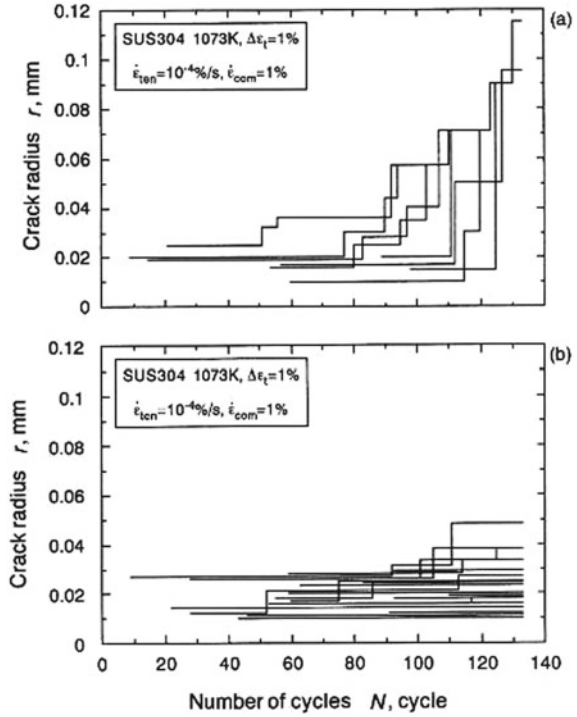
6.4.3.2 Growth Characteristics of Small Inner Crack

Figure 6.19 shows part of the simulated small inner crack propagation curve. Here, the vertical axis is the equivalent crack radius obtained from (6.3). The growing and coalescing cracks are shown in Fig. 6.19a, and the non-propagating cracks are shown in Fig. 6.19b. Small inner cracks propagate at the end of their life, and this rapidly induces crack coalescence, leading to the formation of a main crack, whereas most of the remaining cracks are non-propagating.

Figure 6.20 shows the relationship between the propagation rate dr/dN of an independently propagating crack and the crack radius r , which was determined from the crack propagation curve of the above simulation. Here, cases where cracks coalesce with other cracks are excluded. The crack propagation rate was calculated using the following equation.

$$(dr/dN)_N = (r_{N+\Delta N} - r_N)/\Delta N \quad (6.6)$$

Fig. 6.19 Change in inner crack radius, r , obtained by the simulation [21]



ΔN in the above equation is the number of calculation interval cycles, which is assumed to be 10 cycles. r_N and $r_{N+\Delta N}$ are the crack radii calculated by (6.3) for N cycles and $N + \Delta N$ cycles, and $(dr/dN)_N$ calculated by the above equation corresponds to the crack r_N .

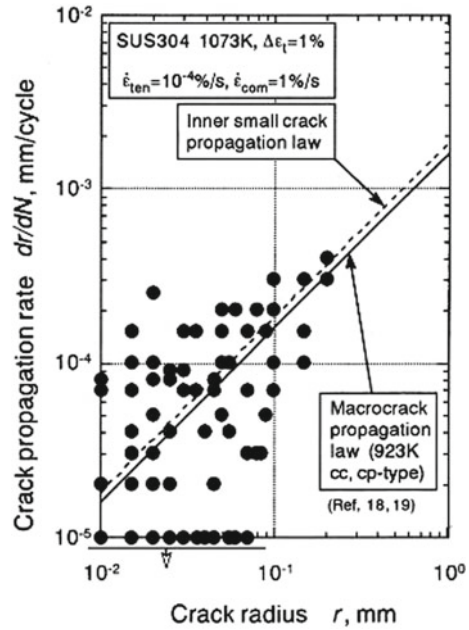
In addition, the solid line in the figure is an extrapolation of the time-dependent macroscopic through-crack propagation law for this material under test conditions (923 K, c-type, cp-type) where no small inner cracks are formed over the entire specimen including the vicinity of the large crack tip. This extrapolation line of the crack propagation law is obtained by the following method. The large crack propagation speed dl/dN can be expressed using the creep J-integral range ΔJ_c as follows⁽¹⁸⁻¹⁰⁾:

$$dl/dN [\text{mm/cycle}] = 2.8 \times 10^{-3} \Delta J_c [\text{kN/m}] \quad (6.7)$$

In addition, the creep J-integral range ΔJ_c of a disc-shaped small crack in an infinite body can be evaluated using the following equation [20].

$$\Delta J_c = \frac{\pi}{6} \left(1 + \frac{3}{n} \right)^{-\frac{1}{2}} r \int_0^{\tau} \sigma \dot{\epsilon}_c dt, \quad (6.8)$$

Fig. 6.20 Relationship between inner crack propagation rate, dr/dN , and crack radius, r , obtained by the simulation [21]



where r is the disk-shaped crack radius, n is the creep index, σ is the nominal stress, $\dot{\epsilon}_c$ is the creep strain rate, and t is the time of the tensile half cycle. Therefore, from the hysteresis loop obtained from the creep-fatigue test, the integral value of the right-hand side of (6.8) is found, and by using (6.7), the extrapolation of the relationship between crack propagation rate and crack radius can be calculated.

In the figure, the propagation rate of small inner cracks varies widely and the cracks are frequently non-propagated. On the other hand, as the crack grows larger, the number of non-propagating cracks decreases, and the variation in crack propagation also decreases, so the small inner crack propagation rate tends to converge to the macrocrack propagation rate curve. This is thought to be due to the relative reduction of local irregularities caused by microstructures as the crack length increases, as in the case of surface small cracks [5-8].

In addition, the dashed line in Fig. 6.20 shows the average propagation rate $(dr/dN)_m$ of small inner cracks in this simulation model. This propagation rate was calculated using the following equation, taking into account the average fracture driving force K related to crack propagation.

$$(dr/dN)_m = \Delta r / (R_m / K \cdot r) = B \cdot r \quad (6.11)$$

$$B = \Delta r \cdot K / R_m \quad (6.12)$$

Here, R_m is the average value of grain boundary fracture resistance ($= 0.5$), and K is the crack propagation driving force ($= 3 \times 10^{-2}$). When the grain boundary facet projection plane is a regular hexagon with an average radius (radius determined by cutting method) of $18 \mu\text{m}$ as shown in Fig. 6.2, Δr is the increase in the equivalent crack radius r determined by (6.3) when a crack with the size of one grain boundary facet propagates. Although Δr changes with r , most of the small inner cracks have a radius of $200 \mu\text{m}$ or less, and the difference in Δr for these cracks is small, so here we consider the crack radius from 18 to $200 \mu\text{m}$. The average value of Δr ($= 30 \mu\text{m}$) was taken as the Δr value.

From Fig. 6.20, it can be seen that the average propagation rate of even small inner cracks agrees with the macrocrack propagation law.

6.4.3.3 Coalescence Effect of Small Inner Cracks

Figure 6.21 shows the relationship between the crack propagation rate dr/dN (increase rate of crack radius) and crack radius r , which was determined by simultaneously considering the growth and coalescence behavior of small inner cracks. The crack propagation rate dr/dN was calculated using (6.6). The propagation rate of small inner crack with coalescence taken into account has a wider variation than that without taking coalescence into account (Fig. 6.20). The thick solid line in the figure is the average value of the propagation rate of small cracks considering coalescence, and the dashed line and thin solid line are the average values of the propagation rate of small inner cracks propagating independently without taking coalescence into account and extrapolation line of macrocrack propagation law, respectively, as in the previous section. The average propagation rate of small inner cracks, taking coalescence into account, is approximately four times higher than the extrapolation line of macrocracks. Considering that the average value of the small inner crack propagation rate without considering coalescence in the previous section matches the extrapolation line of macrocrack propagation law, it can be seen that this acceleration effect is caused by the crack growing while coalescing with other adjacent small cracks.

Figure 6.22 shows some typical examples of small inner crack propagation in order to consider the propagation behavior of individual small inner crack. Even for a single crack, the fast and slow propagation and non-propagation are repeated irregularly. This is thought to be due to the irregularity of fracture resistance and random coalescence of grain boundary facets in all directions adjacent to small inner cracks. Such variations in the crack propagation rate due to repeated acceleration and deceleration are considered to be peculiar to small cracks, since the phenomena do not appear to be as large at longer crack lengths (in the case of macrocrack propagation).

6.4.3.4 Characteristics of Fracture Driving Force F and K

Figure 6.23 shows the crack density n_p on the longitudinal section of the fracture driving force F related to crack initiation and the change in average crack length a_m

Fig. 6.21 Relationship between inner crack propagation rate, dr/dN , and crack radius, r , obtained by the simulation [22]

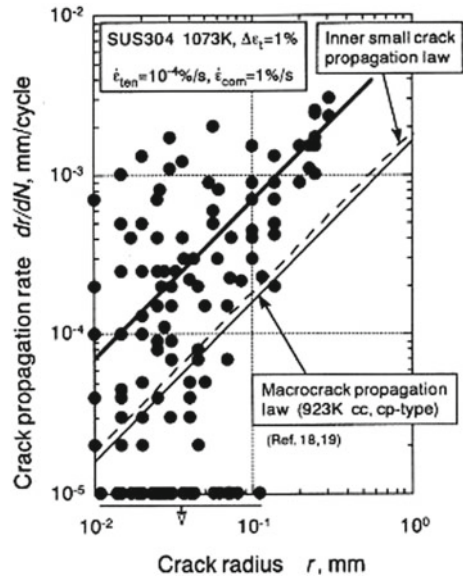
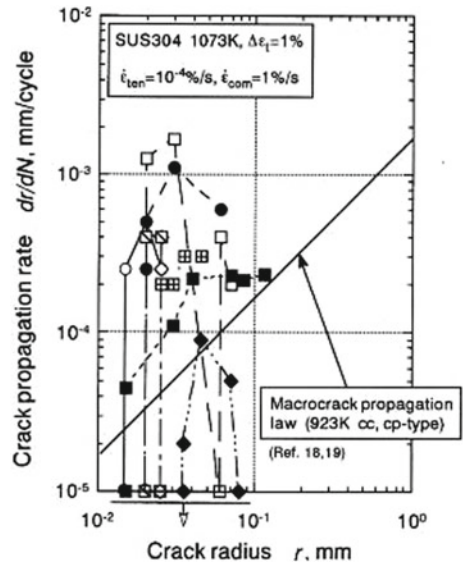
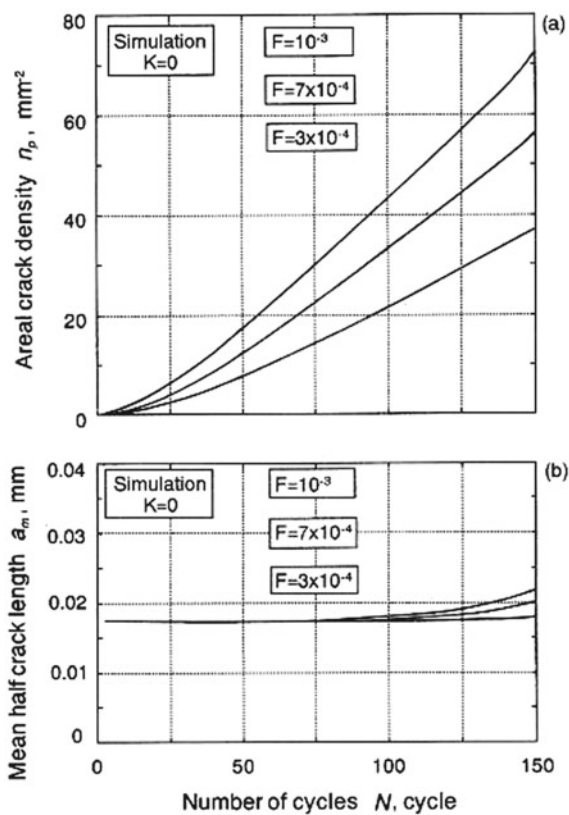


Fig. 6.22 Behavior of inner small cracks propagating obtained by the simulation [22]



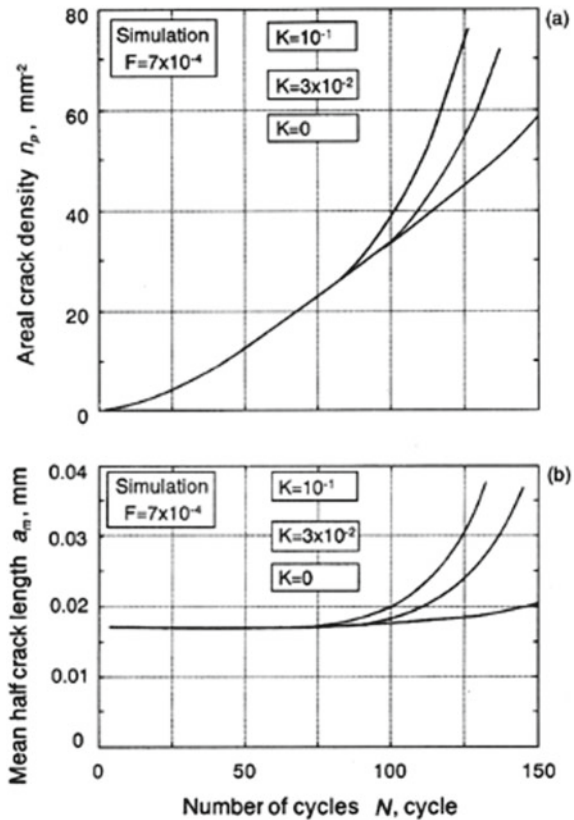
when the value of the fracture driving force K related to crack propagation in (6.2) is fixed. On the other hand, Fig. 6.24 similarly shows their changes due to K when F is fixed. From Figs. 6.23a and 6.24a, it can be seen that the crack density is controlled by F . In Fig. 6.24a, it is observed that as the value of K increases, the crack density tends to increase in the latter half of the life. This is because the cracks grow with

Fig. 6.23 Simulation results ($K = 0$). **a** Change in crack density, n_p . **b** Change in mean crack length, a_m [21]



K without increasing the inner crack density, and the number of cracks that appear on the longitudinal section increases with increasing the size. On the other hand, in Figs. 6.23b and 6.24b, although the initial average crack length coincides with half the average grain boundary length ($18 \mu\text{m}$), the crack length increases as the number of cycles increases. K controls this increase.

Fig. 6.24 Simulation results ($F = 7 \times 10^{-4}$). **a** Change in crack density, n_p . **b** Change in mean crack length, a_m ⁽²¹⁾



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Chapter 7

Effect of Small Inner Cracks on Macrocrack Propagation



Abstract In this chapter, time-dependent and cycle-dependent crack propagation tests were conducted using notched specimens of SUS304 stainless steel in 4 major cases to investigate the effect of small inner cracks introduced by high-temperature cp-type creep-fatigue on macrocrack propagation. For the case of a macrocrack propagation in the zone containing no small crack (Case 1), the macrocrack propagation under cc-type and cp-type conditions is time-dependent, while the macrocrack propagation under pp-type and pc-type conditions is cycle-dependent. For the case of a macrocrack propagation in the zone containing uniformly distributed pre-small cracks (Case 2), the existing small inner cracks accelerate both time-dependent and cycle-dependent macrocrack propagation. For the case of a macrocrack propagation in the process zone with small crack nucleation (Case 3), many small intergranular cracks were observed near the main crack inside the specimen, and the macrocrack propagation rate was approximately 4 times higher than that of the time-dependent crack propagation of case 1. For the case of a macrocrack propagation combined with Case 2 and Case 3 (Case 4), the macrocrack propagation rate is almost equal to that in Case 3 under same conditions.

7.1 Introduction

From the studies in the previous chapter, we found that in creep-fatigue tests of smooth specimens at high temperatures, low tensile strain rates, and high compressive strain rates (cp-type waveform), many small cracks occur inside the material, resulting in a shortened creep-fatigue life. Considering the case where a crack at a notch, which is a stress concentration area, propagates through a region where small inner cracks exist, it is expected that a macrocrack propagation will be greatly affected by frequent coalescence. A large number of researchers have conducted fracture mechanics studies on macroscopic through-crack propagation under high-temperature fatigue conditions, and the fracture mechanics law for the crack propagation rate has also been clarified [1–17]. However, most of the results were obtained under relatively low temperature, high stress, or high strain rate conditions in which

small inner cracks do not initiate. Considering that crack propagation behavior under low-stress, long-period conditions is often a problem in welded joints and stress-concentrated parts of actual high-temperature equipment, it is important to elucidate the influence of small cracks on macrocrack propagation during high-temperature fatigue and creep-fatigue.

As schematically shown in Fig. 7.1, the relationship between small inner cracks and macrocracks can be classified into the following four cases.

Case 1: A macrocrack propagation in the zone containing no small crack.

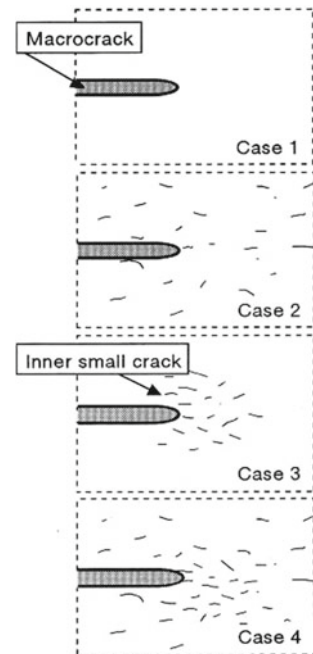
Case 2: A macrocrack propagation in the zone containing uniformly distributed pre-small cracks.

Case 3: A macrocrack propagation in the process zone with small crack nucleation.

Case 4: A macrocrack propagation combined with Case 2 and Case 3.

In this chapter, macrocrack propagation tests for time-dependent creep-fatigue and cycle-dependent fatigue are performed for these four cases, and the influence of small inner cracks is discussed.

Fig. 7.1 Relationship between inner small cracks and a macrocrack. Case 1: A macrocrack propagation in the zone containing no small crack. Case 2: A macrocrack propagation in the zone containing uniformly distributed pre-small cracks. Case 3: A macrocrack propagation in the process zone with small crack nucleation. Case 4: A macrocrack propagation combined with Case 2 and Case 3 [18, 19]



7.2 Fracture Mechanics Law for High-Temperature Fatigue Macrocrack Propagation

7.2.1 Cycle-Dependent Fatigue Crack Propagation

Crack propagation tests on low-alloy steels and austenitic stainless steels have revealed that crack propagation during high-temperature fatigue can be divided into two types: time-dependent creep-dominated crack propagation and cycle-dependent fatigue-dominated crack propagation [1–17]. Among them, cycle-dependent crack propagation is fatigue crack propagation due to blunting and resharping of the crack tip, similar to room temperature fatigue, and in many materials, it is a transgranular fracture in which the main crack propagates inside the grain [1, 2]. As shown in Fig. 7.2a, in the pp-type (a high-frequency waveform in which only non-time-dependent plastic strain occurs in both the tensile and compressive directions and creep strain is almost negligible) and pc-type (waveforms in which creep strain occurs only in the compressive direction), a cycle-dependent crack propagation behavior appears.

According to conventional research based on fracture mechanics such as the J-integral [20–22], in the case of cycle-dependent crack propagation, the propagation rate dl/dN can be calculated using the fatigue J-integral range [6] (cyclic J-integral [23]) ΔJ_f and can be expressed by the following equations [1, 2].

$$dl/dN_N = C_f \Delta J_f^{m_f}. \quad (7.1)$$

Here, C_f and m_f are material constants. The fatigue J-integral range ΔJ_f is defined as the fluctuation range of the J-integral during one cycle as shown in the following equation.

$$\Delta J_f = J_{\max} - J_{\min}. \quad (7.2)$$

Here, $J_{\max}$ and $J_{\min}$ are the maximum and minimum J-integral values during one cycle. If there is a crack closure, $J_{\min}$ is the J-integral corresponding to the load at the closure point, and ΔJ_f is the difference between the J-integral at the closure point and the maximum load point.

The ΔJ_f value for a center cracked plate specimen can be evaluated using the following formula [6, 23].

$$\Delta J_f = \Delta J_e + \Delta J_p \quad (7.3)$$

$$\Delta J_e = (\Delta K_{\text{eff}})^2 / E \quad (\text{plane stress}) \quad (7.4)$$

$$\Delta J_p = 2S_p / A_{\text{net}} \quad (7.5)$$

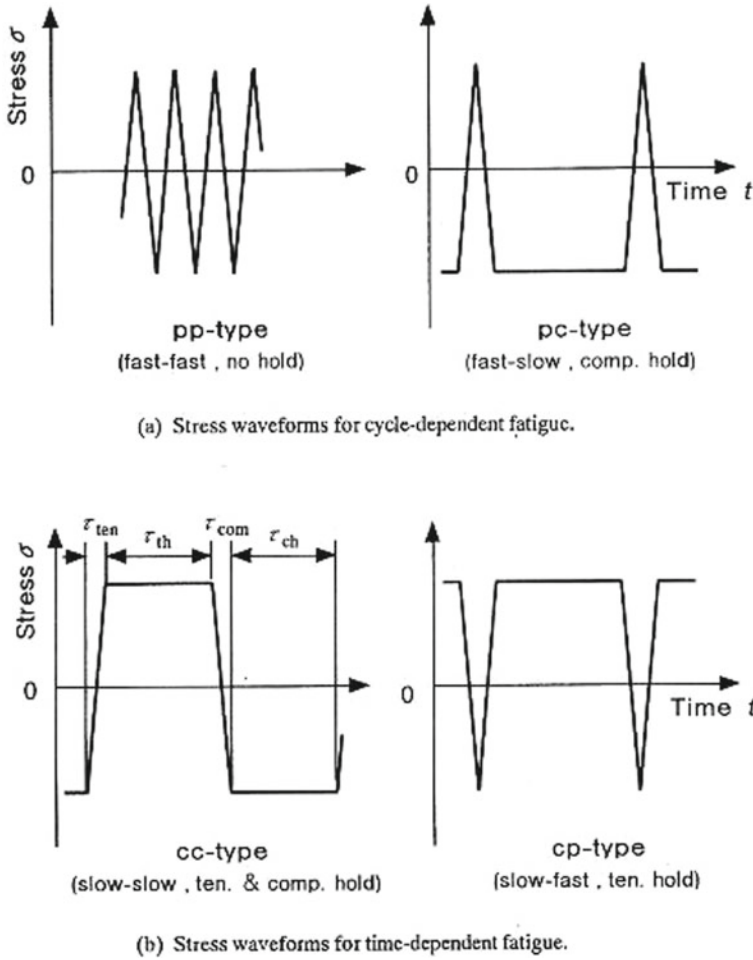


Fig. 7.2 Stress waveforms for high-temperature creep-fatigue crack propagation tests [18]

Here, ΔJ_e is the elastic component of ΔJ_f . ΔJ_p is the plastic component of ΔJ_f . S_p is the shaded area of the hysteresis loop of load P -crack center opening displacement V shown in Fig. 7.3a. ΔK_{eff} is the effective stress intensity factor range. E is Young's modulus. A_{net} is the net cross-sectional area of the specimen.

In the case of high cycle fatigue, the plastic component ΔJ_p of the fatigue J-integral range ΔJ_f becomes 0, and (7.1) agrees with Paris law [24].

$$dl/dN = C(\Delta K_{eff})^m. \quad (7.6)$$

Here, C and m are material constants.

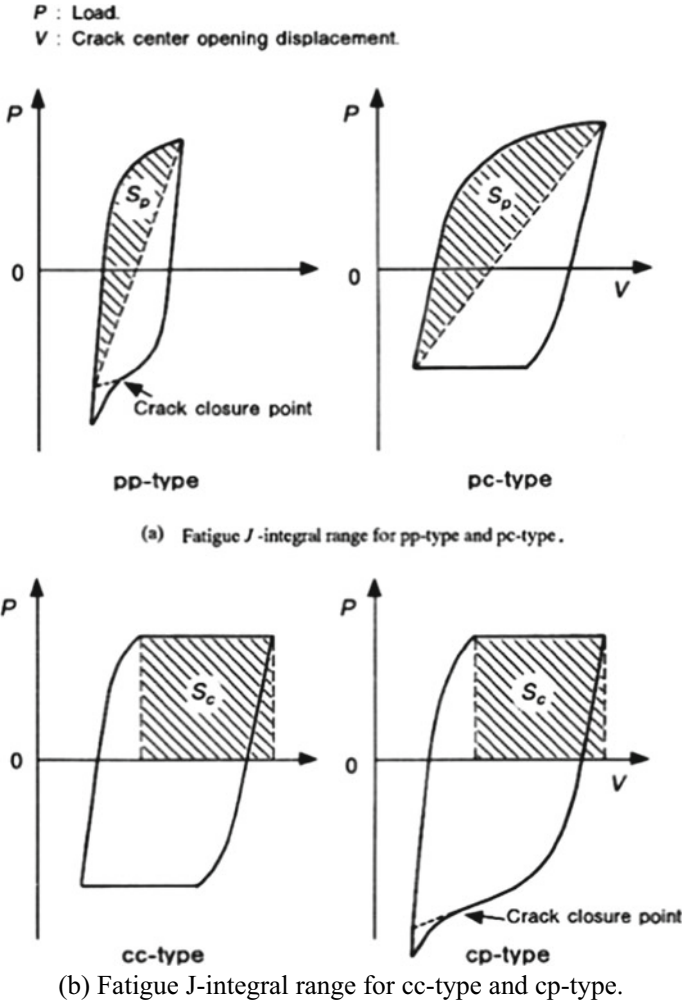


Fig. 7.3 Evaluation method of fatigue J-integral range, ΔJ_f , and creep J-integral range, ΔJ_c , using the hysteresis loops of load and crack center opening displacement [18, 19]

7.2.2 Time-Dependent Crack Propagation

Time-dependent crack propagation is caused by creep strain that accumulates near the crack tip, and in polycrystalline heat-resistant steel, the intergranular cracking often appears and is similar to static creep crack propagation under a constant load [1, 2]. The time-dependent crack propagation is observed when creep strain occurs in tension, as shown in Fig. 7.2b for cc-type (a waveform in which creep strain occurs in both the tensile and compressive directions) and cp-type (a waveform in which creep strain occurs only in the tensile direction).

The time-dependent crack propagation rate dl/dt has a corresponding relationship with the creep J-integral J^* [22] as shown in the following equations [1, 2].

$$dl/dt = C_c (J^*)^{m_c} \quad (C_c, m_c: \text{material constant}). \quad (7.7)$$

Here, the creep J-integral J^* can be defined as the following equation by converting the displacement field in the J-integral into a displacement rate field [22].

$$J^* = \int_{\Gamma} (W^* v_x - T_i \cdot \dot{u}_{i,x}) ds. \quad (7.8)$$

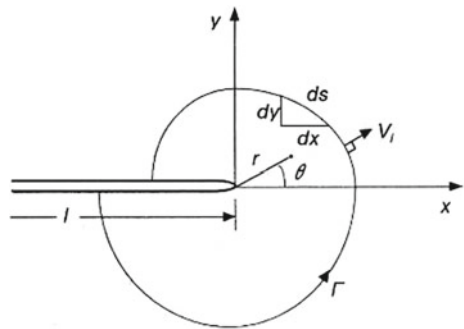
Here, as shown in Fig. 7.4, Γ is the integral path within the material from one point on the bottom surface of the crack to one point on the top surface, and T_i is the component of the stress vector acting on the crack surface $T_i = \sigma_{ij} v_j$, v_j are the components of the outward unit normal vector of the integral path Γ , $\dot{u}_{i,x}$ is the x partial differential of the displacement velocity vector component $\dot{u}_i$, and W^* is the strain rate energy density function. σ_{ij} , $\dot{\epsilon}_{ij}$ are the stress tensor and strain rate tensor, respectively.

$$W^* = \int_0^{\dot{\epsilon}_{ij}} \sigma_{ij} d\dot{\epsilon}_{ij}. \quad (7.9)$$

In many cases, it can be assumed that $m_c = 1$ in (7.7), so this relationship can be obtained by integrating both sides over a tensile half cycle. However, the crack propagation rate dl/dN per unit number of cycles is the creep J-integral range ΔJ_c , and τ is the time of a tensile half cycle. The ΔJ_c value for a center cracked plate specimen can be easily evaluated using the following formula [6, 7].

$$dl/dN = C_c \Delta J_c \quad (7.10)$$

Fig. 7.4 J-integral path and coordinates [19]



$$dl/dN = \int_0^{\tau} (dl/dt) dt \quad (7.11)$$

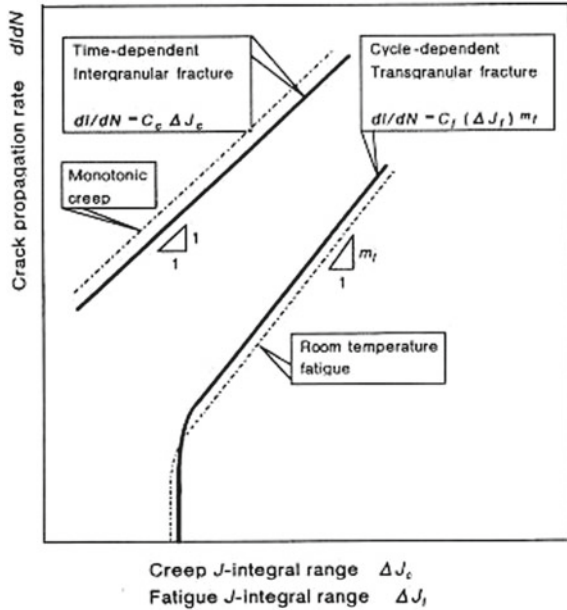
$$\Delta J_c = \int_0^{\tau} J^* dt \quad (7.12)$$

$$\Delta J_c = S_c / A_{\text{net}}. \quad (7.13)$$

Here, S_c is the shaded area of the hysteresis loop between load P and crack opening displacement V shown in Fig. 7.3b, and A_{net} is the net cross-sectional area of the specimen.

Combining the above results, Fig. 7.5 shows a schematic diagram [14] of the $dl/dN - \Delta J_f$ diagram and the $dl/dN - \Delta J_c$ diagram. In general, time-dependent crack propagation exhibits a higher propagation rate than cycle-dependent crack propagation.

Fig. 7.5 Relationship between cycle-dependent fatigue crack propagation rate and fatigue J-integral range, and between time-dependent fatigue crack propagation rate and creep J-integral range [19]



7.2.3 Conditions for Transition from Cycle-Dependent Crack Propagation to Time-Dependent Crack Propagation

The transition conditions from cycle-dependent crack propagation to time-dependent crack propagation or vice versa can be determined by the relationship between the crack propagation rates determined by (7.1) and (7.10) for given ΔJ_f and ΔJ_c values [16]. That is, the following equation is the transition condition (boundary condition).

$$C_c \Delta J_c = C_f \Delta J_f^{m_f}. \tag{7.14}$$

When $C_c \Delta J_c < C_f \Delta J_f^{m_f}$, it becomes a cycle-dependent type, and when $C_c \Delta J_c > C_f \Delta J_f^{m_f}$, it becomes a time-dependent type.

7.3 Test Method

7.3.1 Specimen and Test Equipment

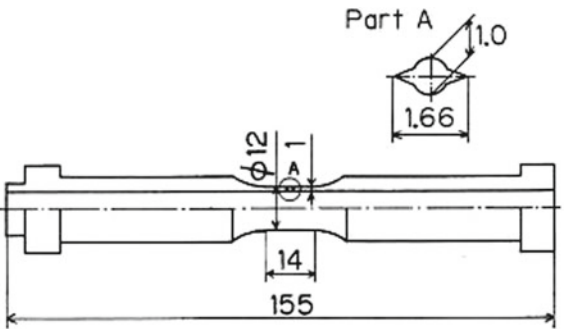
The material is the austenitic stainless steel SUS304 which is same as used in Chap. 3, and its chemical composition and heat treatment conditions are shown in Table 7.1. The hollow thin-walled cylindrical specimen shown in Fig. 7.6 was used for the crack propagation test. A hole with a diameter of 1.0 mm was made in the center of the parallel part of the specimen, and a sharp notch was made with an edge file.

Table 7.1 Chemical composition (wt.%) and conditions of heat treatment [19]

C	Si	Mn	P	S	Cu	Ni	Cr	Mo	Fe
0.06	0.31	1.07	0.31	0.018	0.44	8.68	18.37	0.25	Bal

Heat treatment: 1323 K $\times$ 1 h $\rightarrow$ WQ

Fig. 7.6 Shape and size of notched specimens (dimensions in mm) [19]



The creep-fatigue crack propagation test was conducted using an electric/hydraulic servo high-temperature fatigue testing machine equipped with a high-frequency heater. The crack length was measured using a microscope at any time during the test, and the propagation rate was determined from the slope of the crack propagation curve. In addition, in order to evaluate the value of the J-integral range, the tip of a quartz glass rod of a differential transformer type extensometer was pressed against the stainless steel wire spot-welded at the top and bottom ends of the center of the notch, and the crack center opening displacement V was measured. The P and V hysteresis loops were recorded on an X–Y recorder. In addition, after the test, the area near the crack was polished by approximately 0.5 mm (half of the plate thickness) and the internal appearance of the specimen was observed using an optical microscope.

7.3.2 Test Conditions

The test was conducted in the atmosphere under nominal stress control (load control). The stress waveform (load waveform) used in the test is shown in Fig. 7.2. Note that in the cp-type and pc-type, the shape of the hysteresis loop is different between the tension side and the compression side, and the specimen ratchets as the crack propagates under purely alternating stress. Thus, the compressive load was fine-tuned at any time during the test to prevent ratcheting. The details of the test conditions are shown in Table 7.2 for the four cases mentioned above. Some of the test conditions at 923 K in the table were conducted in the past [10] and were used for comparison with the current test results. The specimens marked 70% N_f , 85% N_f , and Aging under the test conditions of Case 2 and Case 4 in the table refer to the following. Before conducting the crack propagation tests in Case 2 and Case 4, a creep-fatigue test was conducted using the smooth specimen before notching under the conditions shown in Table 7.3 to introduce small inner cracks into the specimen. Figure 7.7 shows an optical micrograph of the center of the thickness of the specimen after the failure. Many small cracks were initiated at the internal grain boundaries of the specimen, and due to their coalescence, the failure occurred at 730 cycles. In Chaps. 4 and 6, cp-type creep-fatigue tests on solid round bar specimens of the same material (strain control, total strain range: 0.7%, tensile strain rate: $10^{-4}\%/s$, compressive strain rate: $1\%/s$, temperature: 1073 K) and numerical simulations were performed. Small cracks began to appear at about 15% of the life, then rapidly increased in number, and at about 75% of the life, the volumetric crack density per unit volume reached 1000 cracks/mm³. The size of cracks is approximately one grain. In addition, after 90% of the life, the density of small inner cracks becomes extremely high, and even a small amount of propagation induces the coalescence of many cracks, causing the inner cracks to grow rapidly, and some of them became macrocrack. Taking these into consideration, we conducted a small inner crack introduction test up to the number of cycles corresponding to 70 and 85% of the life (hereinafter referred to as 70 and 85% pre-damaged specimens, respectively), and then a notch of approximately 2 mm

Table 7.2 Test conditions [19, 25]

(a) Case 1										
Test No	Stress waveform	Max. stress $\sigma_{\max}$ (MPa)	Min. stress $\sigma_{\min}$ (MPa)	τ_{th} (s)	τ_{ch} (s)	τ_{ten} (s)	τ_{com} (s)	Temp. T (K)		
1	cc	166.7	− 166.7	300	300	1	1	923		
2	cc	147.1	− 147.1	480	480	1	1	923		
3	cc	120.0	− 120.0	1000	1000	1	1	923		
4	cc	40.0	− 40.0	1000	1000	1	1	1073		
5	cc	60.0	− 60.0	1000	1000	1	1	1073		
6	cp	166.7	− 191.2 to − 210.9	480	0	2.5	2.5	923		
7	cp	147.1	− 176.5 to − 196.1	480	0	2.5	2.5	923		
8	cp	120.0	− 151.0 to − 169.0	0	0	1000	0.5	923		
9	pp	215.8	− 215.8	0	0	3	3	923		
10	pp	107.9	− 107.9	0	0	0.25	0.25	923		
11	pc	215.8	− 147.1 to − 176.5	0	480	1	1	923		
(b) Case 2										
Test no.	Amount of pre-damage ^a	Stress waveform	Max. stress $\sigma_{\max}$ (MPa)	Min. stress $\sigma_{\min}$ (MPa)	τ_{th} (s)	τ_{ch} (s)	τ_{ten} (s)	τ_{com} (s)	Temp. T (K)	
12	Aging ^b	cp	120	− 136 to − 145	1000	0	0.5	0.5	923	
13	70% N_f	cp	120	− 142 to − 159	1000	0	0.5	0.5	923	
14	85% N_f	cp	120	− 150 to − 163	1000	0	0.5	0.5	923	
15	85% N_f	pp	108	− 108	0	0	0.25	0.25	923	

(continued)

Table 7.2 (continued)

(c) Case 3										
Test no.	Stress waveform	Max. stress $\sigma_{\max}$ (MPa)	Min. stress $\sigma_{\min}$ (MPa)	τ_{th} (s)	τ_{ch} (s)	τ_{ten} (s)	τ_{com} (s)	Temp. T (K)		
16	cp	40.0	− 76.0 to − 91.2	1000	0	1	1	1073		
17	cp	60.0	− 114.0 to − 120.1	1000	0	1	1	1073		
(d) Case 4										
Test no.	Amount of pre-damage ^a	Stress waveform	Max. stress $\sigma_{\max}$ (MPa)	Min. stress $\sigma_{\min}$ (MPa)	τ_{th} (s)	τ_{ch} (s)	τ_{ten} (s)	τ_{com} (s)	Temp. T (K)	
18	85% N_f	cp	40	− 110 to − 118	1000	0	1	1	1073	

^a Number of cycles to failure of smooth specimen: $N_f = 730$

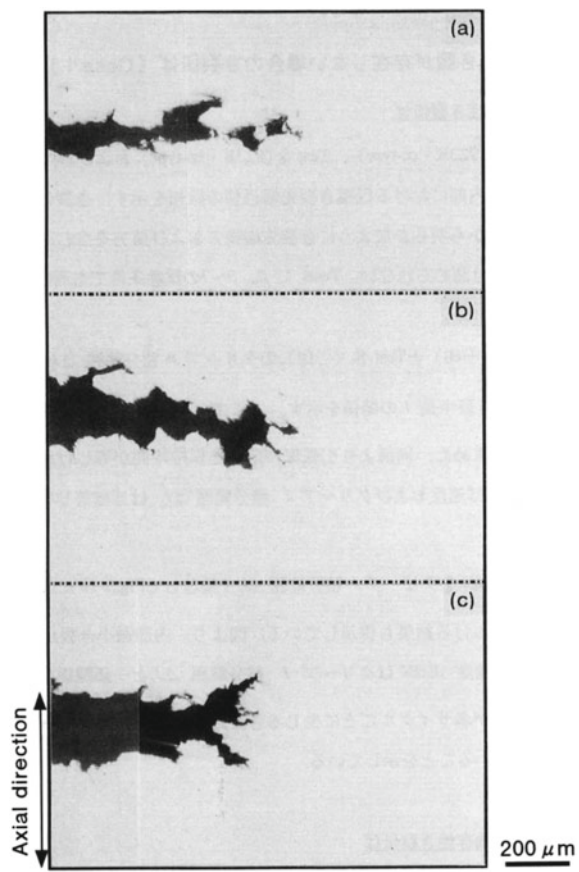
^b Aging time: $\tau = 173$ h

Table 7.3 Test condition

Stress waveform	Max. stress $\sigma_{\max}$ (MPa)	Min. stress $\sigma_{\min}$ (MPa)	τ_{th} (s)	τ_{ch} (s)	τ_{ten} (s)	τ_{com} (s)	Temp. T (K)
cp	45	− 95 to − 114	1000	0	0.5	0.5	1073

Number of cycles to failure of smooth specimen: $N_f = 730$.

Fig. 7.7 Failure morphology near the macrocrack tip in Case 1. **a** cc-type fatigue at 923 K (Test 3). **b** cp-type fatigue at 923 K (Test 8). **c** cc-type fatigue at 1073 K (Test 4) [18]



(Fig. 7.6) was made and subjected to a crack propagation test. In addition, in order to investigate the effect of heating aging added by the small inner crack introduction test, we conducted no-load heating aging (173 h) at 1073 K in the atmosphere for a period of time equivalent to the number of repetitions of 85% pre-damage. The sample was subjected to a crack propagation test.

7.4 Test Results and Discussion

7.4.1 Crack Propagation in the Zone Containing no Small Crack (Case 1)

7.4.1.1 Time-Dependent Crack Propagation

Figure 7.7 shows the morphology near the macrocrack tip inside the specimens of Test 3 (923 K, cc-type), Test 8 (923 K, cp-type), and Test 4 (1073 K, cc-type). Although the crack propagated along the grain boundaries, as is clear from the Figure, no small cracks were observed throughout the specimen. Similar crack propagation patterns are observed under the test conditions of Tests 1, 2, and 5–7.

Figure 7.8 shows the relationship between creep J-integral range ΔJ_c , crack propagation rate dI/dN , and crack half-length l for Test 3 and Test 8. Here, the creep J-integral range ΔJ_c was calculated using (7.13). From the figure, it can be seen that if the tensile holding stress and holding time are equal, the crack propagation speed and creep J-integral range ΔJ_c of cc-type and cp-type are almost equal.

Figure 7.9 shows the crack propagation rate dI/dN organized by the creep J-integral range ΔJ_c . The figure also shows the results under other test conditions. From the figure, it can be shown that when there are no small inner cracks, the crack propagation rate dI/dN has a monovalent relationship with the creep J-integral range ΔJ_c . This indicates that the crack propagation rate is primarily controlled by the tensile creep deformation near the crack tip that occurs every cycle.

7.4.1.2 Cycle-Dependent Crack Propagation

Figure 7.10 shows the morphology near the macrocrack tip inside the specimens of Test 9 (923 K, pp-type) and Test 11 (923 K, pc-type). As is clear from the figure, the macrocrack propagates independently within the grain. Similar crack propagation patterns are observed under other test conditions (Test 10).

Figure 7.11 shows the relationship between fatigue J-integral range ΔJ_f , crack propagation rate dI/dN , and crack half-length l for Test 9 and Test 11. Here, the fatigue J-integral range ΔJ_c was calculated using (7.3). From the figure, under the same conditions (temperature, tensile stress, crack length), the crack propagation rate dI/dN and fatigue J-integral range ΔJ_f of the pc-type are larger than those of the pp-type, and the difference in dI/dN due to the waveform corresponds to the difference in ΔJ_f .

The crack propagation rate dI/dN under the same test conditions is shown in Fig. 7.12, organized by the fatigue J-integral range ΔJ_f . The figure also shows the results of Test 10 and the results previously obtained for the same material [10]. The

Fig. 7.8 Comparison of creep J-integral range and crack propagation rate between cp-type and cc-type at 923 K in Case 1 [18]

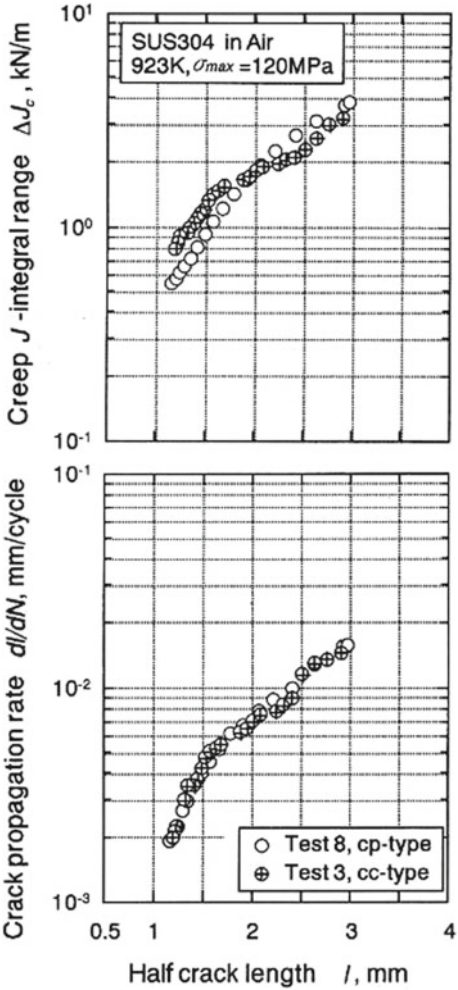


figure shows that the linear relationship expressed by (7.1) is obtained regardless of the difference in waveforms between pp-type and pc-type, and that the fatigue crack propagation rate is determined by the elastoplastic deformation of the tensile half cycle.

Fig. 7.9 Relationship between time-dependent crack propagation rate and creep J-integral range in Case 1 [19]

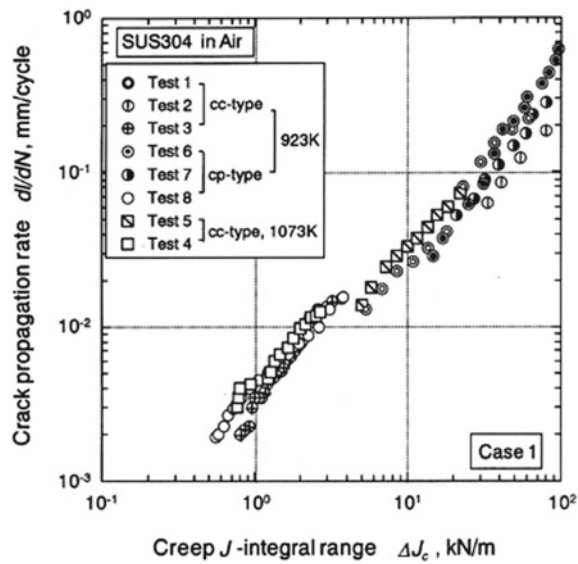


Fig. 7.10 Failure morphology near the macrocrack tip in Case 1. **a** pp-type fatigue at 923 K (Test 9). **b** pc-type fatigue at 923 K (Test 11) [18]

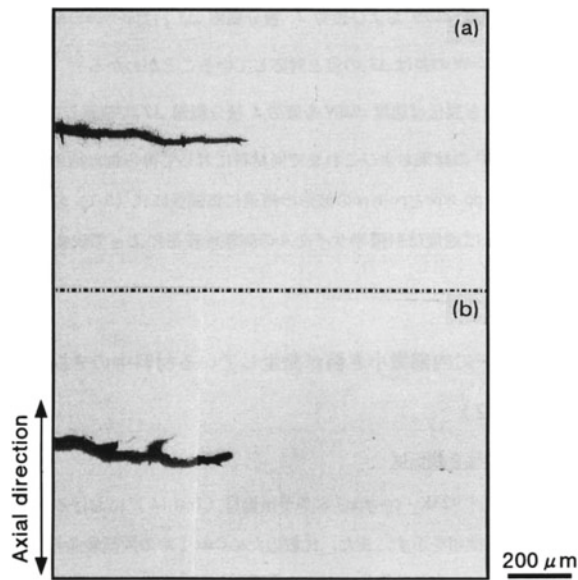
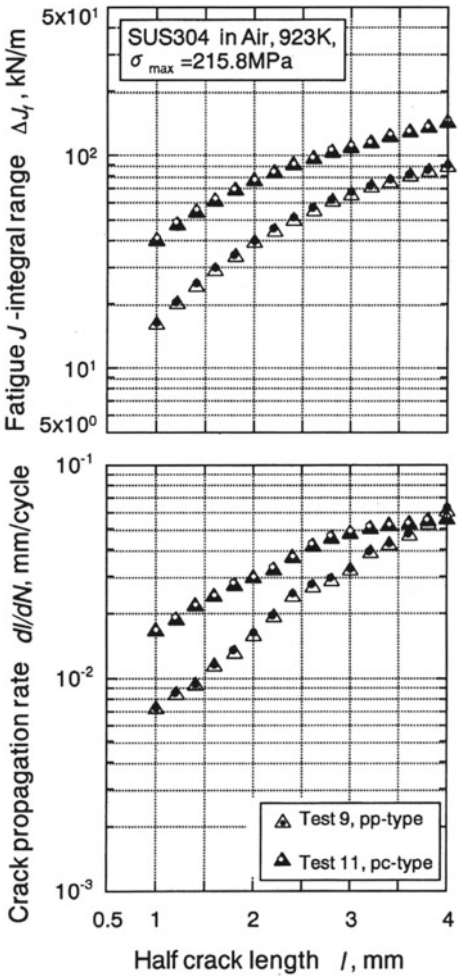


Fig. 7.11 Comparison of fatigue J-integral range, and crack propagation rate between pp-type and pc-type at 923 K in Case 1 [18]



7.4.2 Crack Propagation in the Zone Containing Uniformly Distributed Pre-Small Cracks (Case 2)

7.4.2.1 Time-Dependent Crack Propagation

Figure 7.13a shows the morphology near the macrocrack tip near the center of the plate thickness in a 923 K, cp-type, 85% pre-damaged specimen (Test 14). For comparison, a photograph of the vicinity of the macrocrack tip of the undamaged specimen under the same test conditions (Test 3) in Case 1 is also shown in Fig. 7.13b.

All macrocracks propagate along grain boundaries, but in the pre-damaged specimen there are many small intergranular cracks, and the macrocracks grow by repeatedly coalescing with the small intergranular cracks distributed forward.

Fig. 7.12 The sample was subjected Relationship between cycle-dependent crack propagation rate and fatigue J-integral range in Case 1 [18]

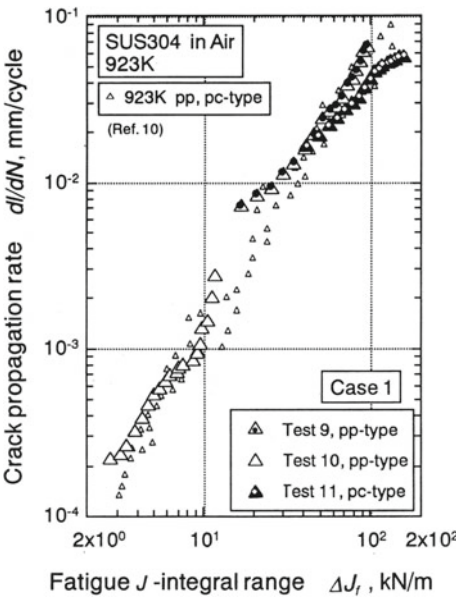


Fig. 7.13 Failure morphology near the macrocrack tip in cp-type fatigue at 923 K. **a** Case 2 (Test 14). **b** Case 1 (Test 8) [18]

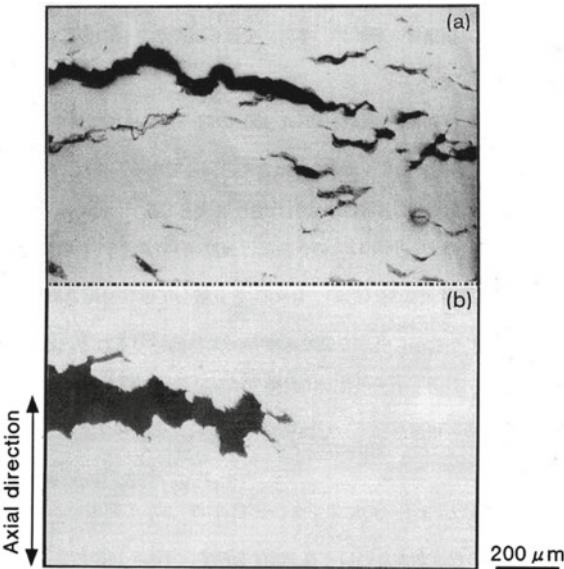
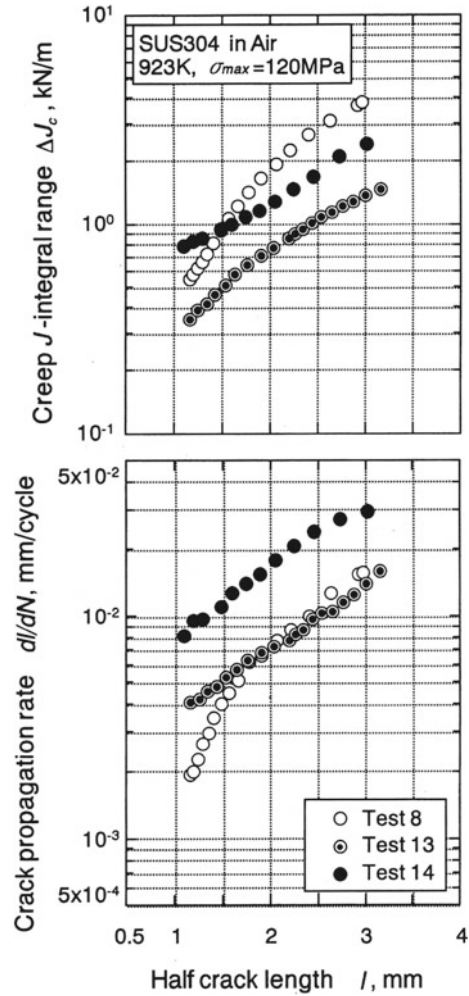


Figure 7.14 shows the changes in creep J-integral range ΔJ_c and crack propagation rate d/dN for different pre-damaged specimens. The figure shows that ΔJ_c for the same crack length is smaller than that for the undamaged specimen with no small inner cracks (Test 8 of Case 1). This is due to the cyclic hardening of the material by

Fig. 7.14 Effect of inner small cracks on creep J-integral range and crack propagation rate in cp-type fatigue at 923 K in Case 2 [18]



pre-creep-fatigue at 1073 K. On the other hand, small cracks uniformly distributed in the material promote the deformation of the entire material and cause softening due to damage, so pre-damaged material has the effect of increasing the ΔJ_c value. Since both effects are included in the crack center opening displacement in (7.12), ΔJ_c of the pre-damaged specimen in Fig. 7.14 is the result of the superposition of the effects of cyclic hardening and softening due to small cracks. Therefore, it is considered that in the 70% damaged specimen (Test 13), the ΔJ_c value decreased due to hardening, but in the 85% damaged specimen (Test 14), the ΔJ_c value increased due to the presence of multiple small cracks. From the lower part of Fig. 7.14, the ΔJ_c value for the same crack length for the 70% pre-damaged specimen (Test 13) is smaller than that for the virgin material, but the dl/dN is almost the same. In addition, the

propagation rate of the 85% pre-damaged specimen (Test 14) is approximately three times faster than that of the undamaged material.

Figure 7.15 shows the relationship between crack propagation rate dI/dN , J-integral range ΔJ_c , and crack length I for aged specimen (Test 12) in comparison with unaged specimen (Test 8). The ΔJ_c of the aged specimen is slightly larger value after crack propagation begins, but it is almost the same as that of the unaged material, and it is clear that the decrease in ΔJ_c of the pre-damaged specimen is not due to heating aging. Furthermore, there is no difference in the crack propagation rate between the two specimens, and heating aging does not have a large effect on crack propagation.

Figure 7.16 shows the relationship between dI/dN - ΔJ_c by combining the results of Figs. 7.14 and 7.15. For the same ΔJ_c , the greater the degree of pre-damage, the higher the dI/dN . Considering the results in Fig. 7.13, it can be understood that the

Fig. 7.15 Effect of aging at 1073 K for 173 h on creep J-integral range and crack propagation rate in cp-type fatigue at 923 K [18, 26]

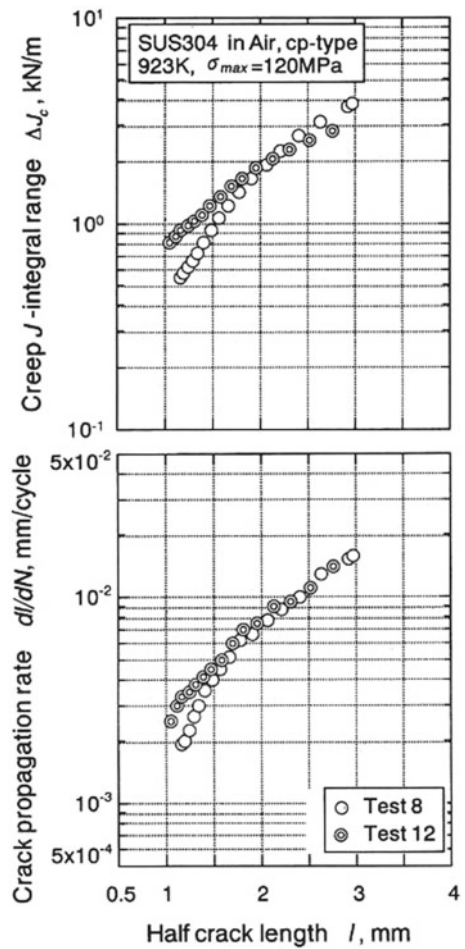
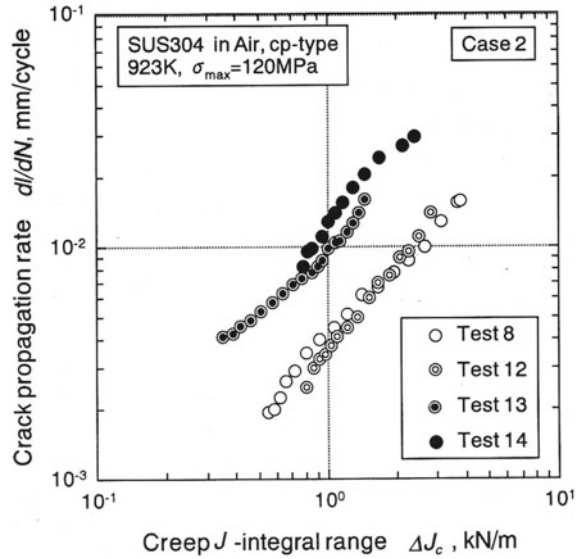


Fig. 7.16 Relationship between crack propagation rate and creep J-integral range in Case 2 [18]



presence of many small intergranular cracks in front of the macrocrack reduces crack propagation resistance and accelerates crack propagation. Furthermore, as mentioned above, even at the same temperature of 923 K, small cracks do not initiate inside the specimen under cc-type fatigue. In the past, macroscopic crack propagation tests were conducted under these conditions after introducing creep-fatigue pre-damage equivalent to 80% of the failure life and it was reported that there was no effect of pre-creep-fatigue on the dl/dN - ΔJ_c relationship [15, 16]. This result also suggests that the acceleration of the propagation rate of the pre-damaged specimen in Fig. 7.16 is caused by small inner cracks.

7.4.2.2 Cycle-Dependent Crack Propagation

Figure 7.17 shows the comparison of macrocrack tip morphology near the center of the thickness of the 85% pre-damaged specimen (Test 15) with that of the undamaged specimen (Test 10). In the undamaged material, as clarified in Sect. 7.4.1.2, no small cracks or secondary cracks were observed, and the morphology of transgranular fracture was similar to room temperature fatigue cracking. On the other hand, macrocracks in pre-damaged specimens in which small inner cracks have been previously introduced by creep-fatigue are dominated by intergranular fractures, although some of them are transgranular fractures. In addition, many small cracks exist around the macrocracks. In other words, the macrocrack propagates through the small intergranular cracks introduced by pre-creep-fatigue.

Fig. 7.17 Failure morphology near the macrocrack tip in pp-type fatigue at 923 K. **a** Case 2 (Test 15). **b** Case 1 (Test 10) [18]

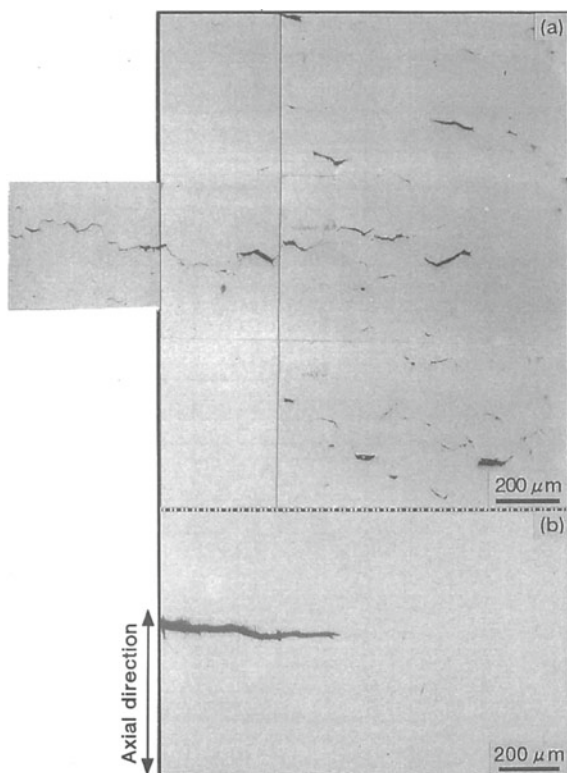
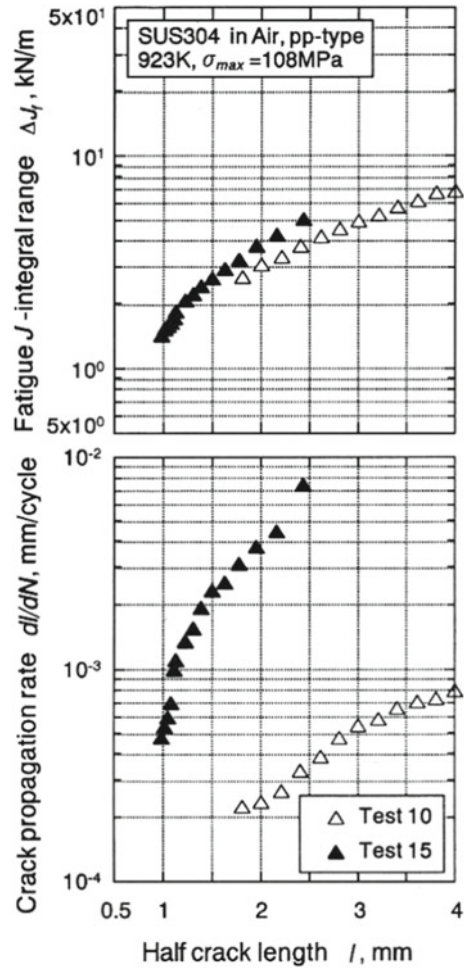


Figure 7.18 shows the influence of small inner crack damage on the fatigue J-integral range ΔJ_f and the crack propagation rate dI/dN . The ΔJ_f of the 85% pre-damaged specimen (Test 15) is slightly larger than that of the undamaged specimen (Test 10), but no major difference is observed. In contrast, it can be seen that dI/dN is accelerated by more than an order of magnitude due to pre-damage.

When the crack propagation rate dI/dN is expressed as ΔJ_f , there is a difference of about 10 times in the same relationship between the two materials, as shown in Fig. 7.19. In other words, the pre-creep damage of small inner cracks introduced by the 1073 K cp-type reduces the crack propagation resistance without changing the deformation resistance (crack propagation driving force) and increases the cycle-dependent fatigue damage. Therefore, this should be taken into account when the cycle-dependent fatigue crack propagation behavior is predicted in cases where creep damage is preceded by cavities or small intergranular cracks inside the material, or where there are many crack-like defects inside the material.

It has already been reported that the pre-damage introduced by 923 K cc-type fatigue, which does not initiate small cracks inside the specimen, has little effect on the subsequent cycle-dependent fatigue crack propagation [15, 16]. This result also

Fig. 7.18 Effect of inner small cracks on fatigue J-integral range and crack propagation rate in pp-type fatigue at 923 K in Case 2 [18, 26]



suggests that the acceleration of the propagation rate of the pre-damaged specimen in Fig. 7.19 is caused by small inner cracks.

7.4.3 Macrocrack Propagation in the Process Zone with Small Crack Nucleation (Case 3)

Figure 7.20a shows the morphology near the macrocrack tip inside the specimen of Test 16 (1073 K, cp-type). As is clear from the figure, cracks propagate through grain boundaries, and many small cracks exist near macrocracks. In other words, in the 1073 K cp-type, small intergranular cracks were initiated throughout the macrocrack

Fig. 7.19 Relationship between crack propagation rate and fatigue J-integral range in Case 2 [18]

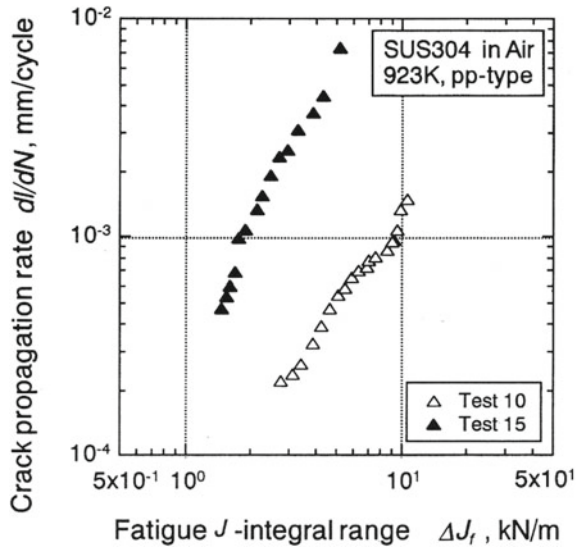
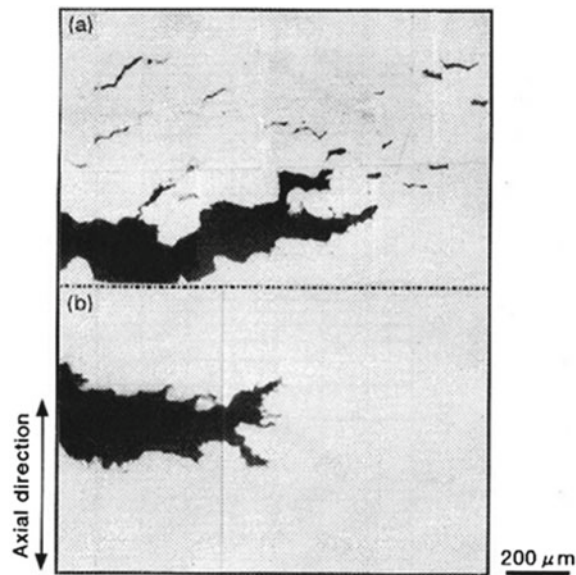


Fig. 7.20 Failure morphology near the macrocrack tip in the cp-type and cc-type fatigue at 1073 K. **a** Case 3 (Test 16). **b** Case 1 (Test 4) [18]



front, and the main crack is considered to have grown by repeatedly coalescing with these small cracks. A similar propagation pattern is observed in Test 17. On the other hand, for comparison, Fig. 7.20b shows the morphology of the macrocrack tip inside the specimen of Test 4 (1073 K cc-type) in Case 1 mentioned above. Although this is an intergranular crack, no small cracks were observed near the crack tip as shown in Fig. 7.20a. The difference between c-type and cp-type is expected to be due to

the suppression and healing effect of compression creep on small inner cracks, as discussed in the next chapter.

Figure 7.21 shows the relationship between creep J-integral range ΔJ_c , crack propagation rate dI/dN , and crack half-length I for Test 16 compared to Test 4. For the same crack length, the crack propagation rate of cp-type is about twice as fast as that of cc-type, but the creep J-integral range is about twice as small.

The propagation rate is organized in terms of the creep J-integral range and is shown in Fig. 7.22. The same figure also shows the results of Test 17 for Case 3 and Test 5 for Case 1. As can be seen from the figure, dI/dN - ΔJ_c of Case 3 is significantly different from that of Case 1. In other words, for the same creep J-integral range ΔJ_c , the crack propagation rate in Case 3 (1073 K · cp-type) is approximately 4 times higher than that in Case 1 (1073 K, cc-type; 923 K, cc, cp-type).

Fig. 7.21 Effect of inner small cracks on creep J-integral range and crack propagation rate in cp-type fatigue at 1073 K in Case 3 [18]

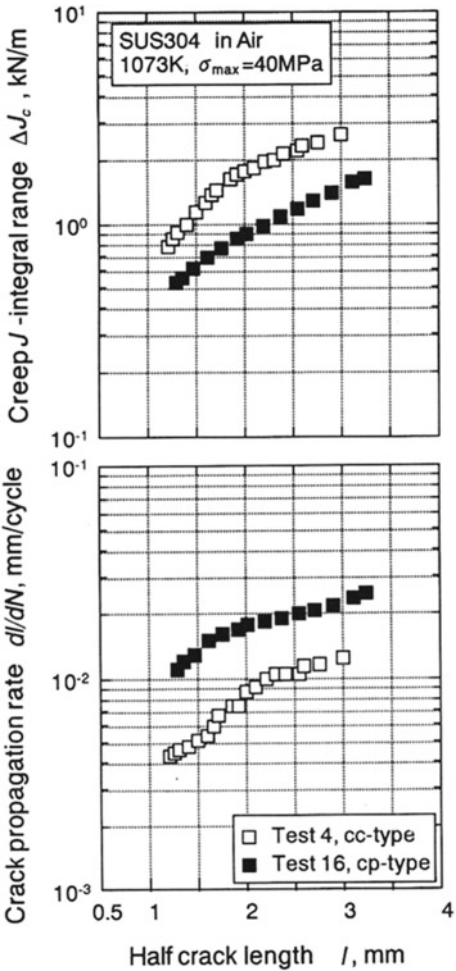
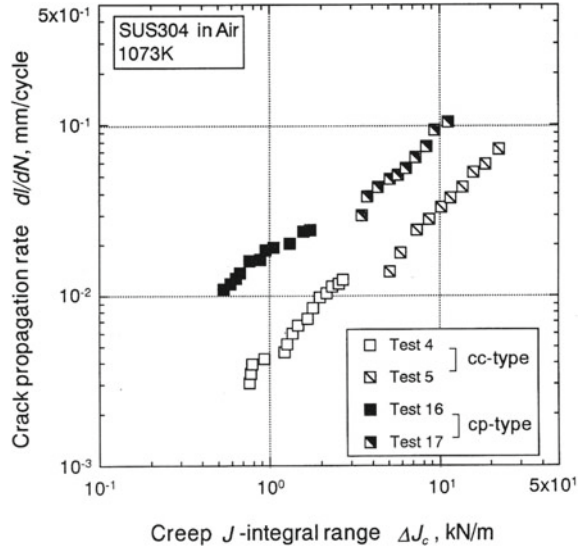


Fig. 7.22 Relationship between crack propagation rate and creep J-integral range in Case 3 [18]



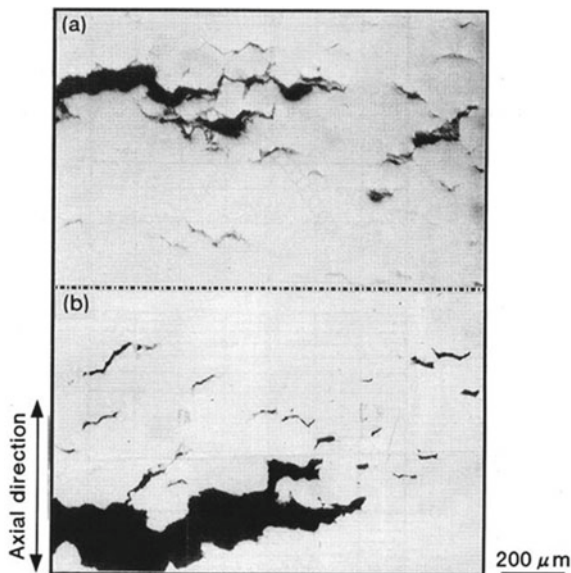
7.4.4 Crack Propagation Combined with Case 2 and Case 3 (Case 4)

Figure 7.23a shows the morphology of the macrocrack tip near the center of the thickness of the 85% pre-damaged specimen (Test 18) at 1073 K, cp-type. Macrocracks propagate along grain boundaries. The large number of small cracks observed at the tip of the macrocrack and near the crack surface indicates that the macrocrack grew while repeatedly merging with the small intergranular cracks distributed in the crack front. For comparison, Fig. 7.23b also shows a photograph of the crack in Case 3 and Test 16, which propagated while forming small inner cracks near the crack tip under the same test conditions. The same point as in Fig. 7.23a is that small intergranular cracks are observed near the crack tip, but their density is slightly lower.

Figure 7.24 shows the ΔJ_c and dI/dN for both specimens. Both ΔJ_c and dI/dN are smaller for the 85% pre-damaged material (Case 4, Test 18) compared to the undamaged material (Case 3, Test 16). However, as the crack length is larger, there is a tendency for the difference from the virgin material to become smaller.

Figure 7.25 shows the relationship between dI/dN and ΔJ_c at 1073 K for both specimens. The relationship between the crack propagation rate and ΔJ_c in Case 4 (Test 18) is almost the same as that in Case 3 (Test 16). However, as mentioned above, the results of Case 3 show higher propagation rate than those of Case 1 for the same ΔJ_c value. In the crack propagation in Case 4, it is considered that the effects of the small cracks initiated during macrocrack propagation are compounded by the small cracks introduced in advance by creep-fatigue. However, as shown in Fig. 7.25, the results for the pre-damaged and undamaged specimens at 1073 K,

Fig. 7.23 Failure morphology near the macrocrack tip in cp-type fatigue at 1073 K. **a** Case 4 (Test 18). **b** Case 3 (Test 16) [19]



cp-type are consistent. This shows that the effect saturates when the density of small cracks exceeds a certain level.

Figure 7.26 summarizes the above-mentioned Figs. 7.9, 7.12, 7.16, 7.19, 7.22, and 7.25. From the figure, it can be seen that Cases 3 and 4 have upper limits and the dI/dN is about 4 times higher than that of Case 1, where there are no small inner cracks. Furthermore, if larger pre-creep-fatigue damage is applied, large cracks will be formed due to the rapid coalescence of small inner cracks, as revealed in Chaps. 4 and 6, making it impossible to obtain a stable propagation curve. It is considered inappropriate to treat these as macrocrack propagation.

Fig. 7.24 Effect of inner small cracks on fatigue J-integral range and crack propagation rate in cp-type fatigue at 1073 K in Case 4 [18, 26]

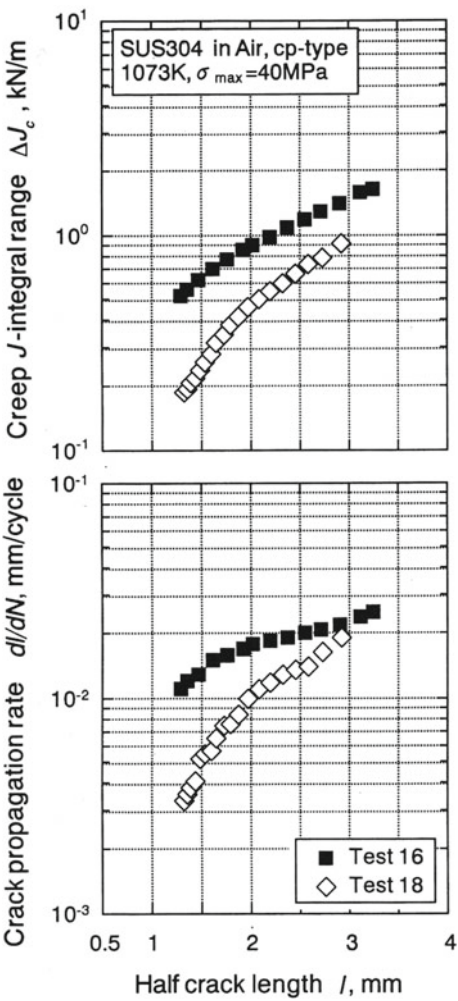


Fig. 7.25 Relationship between crack propagation rate and creep J-integral range in Case 4 [18, 19]

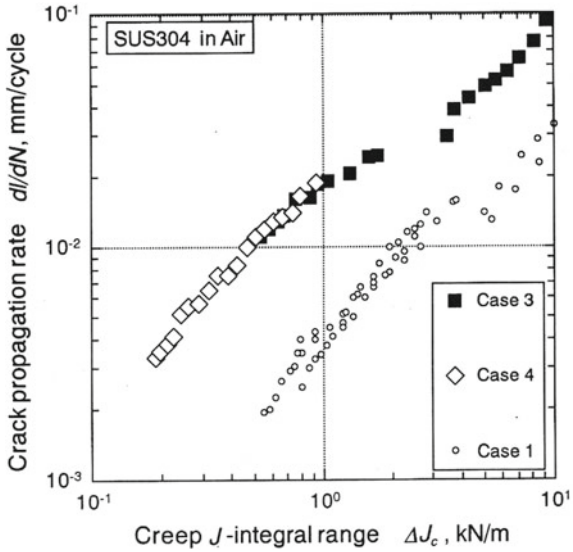
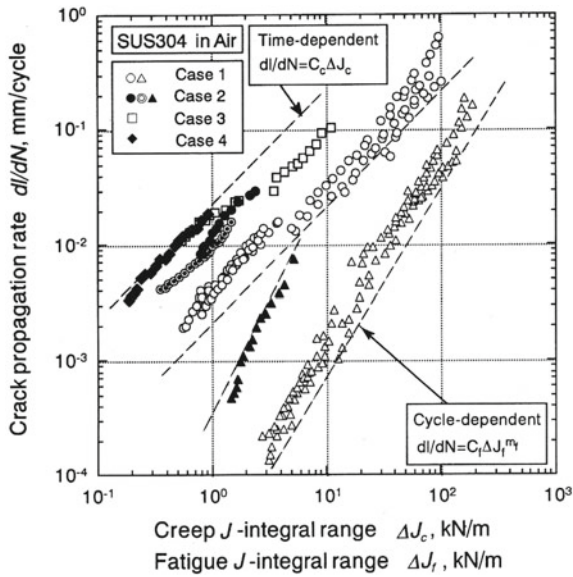


Fig. 7.26 Effect of inner small cracks on the relationship between crack propagation rate and fatigue J-integral range for cycle-dependent fatigue and between crack propagation rate and creep J-integral range for time-dependent fatigue [19, 26]



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Chapter 8

Annihilation and Healing of Small Inner Cracks and Extension of Fatigue Life



Abstract In this chapter, a damage healing due to compressive creep was examined by applying pc-type fatigue loading to Cr–Mo–V forged steel specimens in which cavities and small cracks were introduced to the inside by cp-type creep-fatigue tests. In the slow-tensile fast-compression (cp-type) creep-fatigue tests at high temperature, many small cracks and cavities initiate inside the specimen, and at the end of the life, they coalesce and grow, leading to rapid fracture. In the pc-type fatigue tests, small cracks and cavities do not initiate inside the specimen, and the cracks on the specimen surface grow stably and lead to fracture. After introducing damage to the specimen by loading it to 3/4 of the life of cp-type creep-fatigue, applying pc-type fatigue loading causes the small inner cracks and cavities to shrink and disappear. In other words, pc-type fatigue loading has the effect of healing the damage caused by cp-type fatigue. This healing effect appears even with a small number of cycles (5 cycles) of pc-type fatigue loading. Materials in which small inner cracks and cavities are annihilated and healed by pc-type fatigue loading exhibit a creep-fatigue life equivalent to that of virgin materials. The above-mentioned healing effect is thought to be due to the coalescence of the crack surfaces during compressive creep and the shrinkage of the cavity due to grain boundary diffusion under compressive stress.

8.1 Introduction

Under conditions of high temperature and low tensile stress (or low tensile strain rate), many cavities and small cracks initiate on the grain boundaries inside heat-resistant steel, and they coalesce and grow, resulting in fracture. It was pointed out in Chap. 3 that fatigue life is the shortest in this type of fracture mechanism. On the other hand, even at the same high temperature and low tensile strain rate, when the compressive strain rate is low, no small inner cracks or cavities are observed, and the life tends to be much longer. In addition, as was clarified in Chap. 7, during 1073 K, cp-type ($\varepsilon_{\text{ten}} = 10^{-4}\%/s$, $\varepsilon_{\text{com}} = 1\%/s$) fatigue of SUS304, many cracks initiated near the tip of the macrocrack and the macrocrack propagates while coalescing with the small cracks. In contrast, at the same temperature and cc-type ($\varepsilon_{\text{ten}} = 10^{-4}\%/s$, ε_{com}

$= 10^{-4}\%/s$), no small inner cracks are observed, and the macrocrack propagation rate is about four times lower than that of the cp-type for the same creep J-integral range ΔJ_c value. These facts suggest that compression creep has the effect of annihilating small inner cracks. In addition, its disappearance is considered to be one of the causes of the strain waveform effects [1–3] peculiar to high-temperature creep-fatigue, as mentioned in the introduction. Up until now, attempts have been made to annihilate creep cavities by methods such as high-temperature solution treatment and heating under hydrostatic pressure [4–8], but it seems impossible to annihilate completely the creep cavities and small cracks and to improve the creep fracture life by simple heating. No systematic study has been conducted on the annihilation and healing of small inner cracks due to compression creep. In addition, research has recently begun on intelligent materials with damage self-healing capability [9]. However, it is extremely difficult to develop materials with sufficient functionality as high-temperature structural materials and with self-healing capabilities. In this situation, it is necessary to artificially heal the damage by taking some appropriate measures.

In this chapter, we conducted low-speed tension high-speed compression (cp-type) creep-fatigue tests using Cr–Mo–V forged steel for turbine rotors and performed the above-mentioned cp-type creep-fatigue tests as basic research on the healing of damage caused by small inner cracks. We conducted an observation of the phenomenon in which small inner cracks caused cp-type creep-fatigue disappear when subjected to cyclic high-speed tension low-speed compression (pc-type) strain.

8.2 Test Method

8.2.1 Material and Specimen

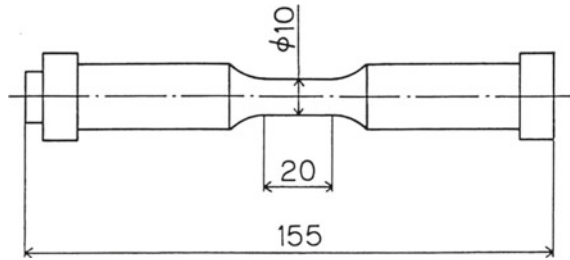
The material was 1Cr–1Mo–1/4V forged steel cut from an unused turbine rotor, and the same material was used in the observation of small inner crack initiation and propagation due to creep-fatigue in Chap. 4. Its chemical composition and heat treatment conditions are listed in Table 8.1. For the creep-fatigue test, a solid round bar smooth specimen with a diameter of 10 mm and a parallel part length of 20 mm as shown in Fig. 8.1 was used. Before the test, the specimen was polished with emery paper up to #3000 to remove the small scratches on the surface.

Table 8.1 Chemical composition (wt.%) and heat treatment

C	Si	Mn	P	S	Cu	Ni	Cr	Mo	V	Fe
0.28	0.24	0.75	0.006	0.004	0.03	0.48	1.13	1.19	0.25	Bal

Heat treatment: 1243 K $\times$ 20 h $\rightarrow$ AC; 948 K $\times$ 38 h $\rightarrow$ FC

Fig. 8.1 Shape and size of specimen [2]



8.2.2 Test Conditions

The test was conducted in the atmosphere at 923 K (650 °C) under strain control using an electric/hydraulic servo fatigue testing machine equipped with a high-frequency heater. Two types of strain waveforms were used: pc-type with the total strain range of 1.0%, high-speed tension (tensile strain rate 1%/s), low-speed compression (compression strain rate $3 \times 10^{-4}\%$ /s) waveform, and cp-type with low-speed tension (tensile strain rate $3 \times 10^{-4}\%$ /s) and high-speed compression (compressive strain rate 1%/s) waveform as shown in Fig. 8.2. The test conditions are listed in Table 8.2. In Tests 1 and 2, pc-type and cp-type strain waveforms were applied, respectively, until the failure life N_f (the number of cycles at which the tensile stress decreased to 3/4 of the steady state). The life (N_f)_{pc} of Test 1 (pc-type) was 606 cycles, and the life (N_f)_{cp} of Test 2 (cp-type) was 86 cycles. In Test 3, after applying a cp-type strain waveform up to 3/4 of (N_f)_{cp}, the test was interrupted and a pre-damaged specimen was obtained with creep cavities and small cracks introduced inside the specimen. In Tests 4–6, following the pre-damage test of Test 3, pc-type tests were conducted to observe the annihilation of cavities and cracks. In Test 7, after the pre-damage test of Test 3, the pc-type was inserted 30 cycles, surface cracks were removed by polishing, and the subsequent cp-type creep-fatigue was performed. In other tests, surface small cracks were not removed. For Tests 1, 2, 6, and 7, the specimens were tested until the failure life, and for the other tests, the specimens were subjected to crack observation after a specified number of creep-fatigue cycles.

8.2.3 Observation Method

Cavities and small cracks inside the specimen were observed using an optical microscope and a scanning electron microscope after the specimen was cut vertically and subjected to buffing and electrolytic polishing. Electrolytic polishing was performed using Jacquet solution (acetic anhydride: perchloric acid = 4:1) on an area of 50 mm² at a constant current of 0.7 A for 50 s. In addition, we performed electrolytic polishing on unused 1Cr–1M0–1/4V forged steel under the same conditions and confirmed that no pits were formed. The density and length distribution of cavities

Fig. 8.2 Strain waveforms
[2]

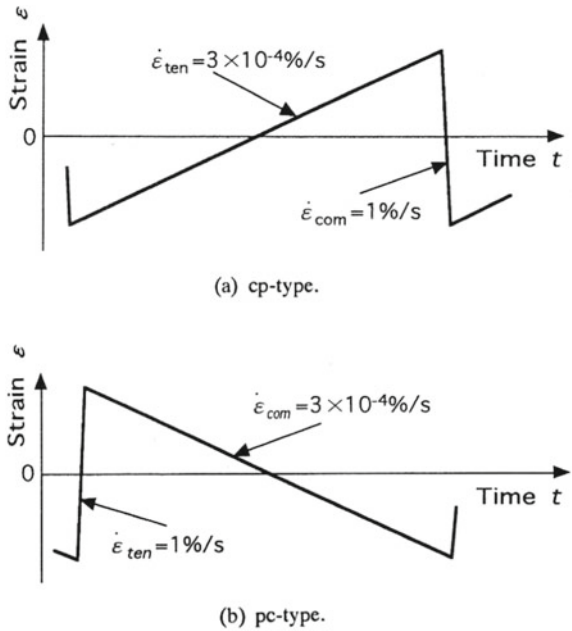


Table 8.2 Test conditions ($T = 923\text{ K}$, $\Delta\epsilon_t = 1\%$)

Test no.	Type of creep-fatigue	
Test 1	pc-type 606 cycles	Failure test
Test 2	cp-type 86 cycles	Failure test
Test 3	cp-type 64 cycles	Interrupted test
Test 4	cp-type 64 cycles + pc-type 5 cycles	Interrupted test
Test 5	cp-type 64 cycles + pc-type 30 cycles	Interrupted test
Test 6	cp-type 64 cycles + pc-type 340 cycles	Failure test
Test 7	cp-type 64 cycles + pc-type 30 cycles + cp-type 83 cycles Surface was polished after the pc-type fatigue	Failure test

and small cracks were measured by photographing an approximately 2 mm^2 area of the electrolytically polished surface (50 mm^2), including the central axis of the specimen, and magnifying it 400 times. Preliminary observations confirmed that cavities and cracks were uniformly distributed inside the specimen.

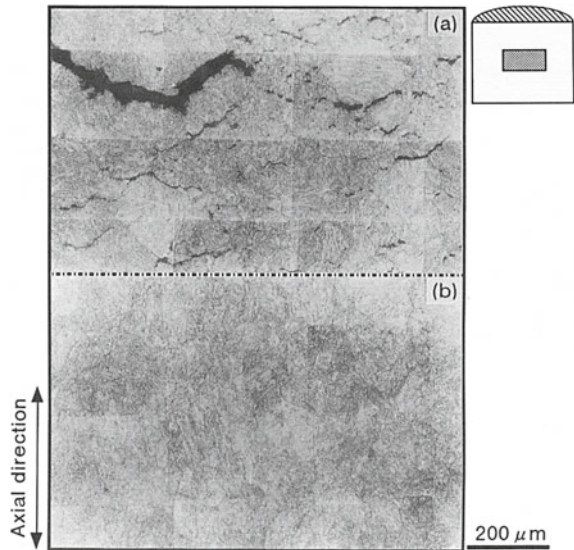
8.3 Test Results and Discussion

8.3.1 Fracture Morphology and Differences Between Cp-Type and Pc-Type Fatigue

The failure life $(N_f)_{cp}$ in low-speed tension high-speed compression (cp-type, Test 2) was 86 cycles, whereas the failure life $(N_f)_{pc}$ in high-speed tension low-speed compression (pc-type, Test 1) was 606 cycles. In the cp-type, many small cracks initiate on the internal grain boundaries, as shown in Fig. 8.3a. The main crack that caused the failure was also formed internally. A photograph of the entire main crack on the longitudinal section is shown in Fig. 8.4a. From the photographs, it is clear that many small cracks coalesce to form a main crack. This is the typical internal cracking type fracture described in Chap. 3. Figure 8.5a shows the change in maximum and minimum stress over time. The figure shows that at the end of the life of Test 2 (cp-type), the stress begins to decrease and the fracture occurs in a short time, but many small cracks that have initiated inside the specimen rapidly coalesce and grow to form a main crack at the end of the life, leading to failure.

In the pc-type (Test 1), as shown in Fig. 8.3b, no small cracks were observed inside the specimen, and as is clear in Fig. 8.4b, the failure was caused by cracks that were initiated and grew on surface, and the failure morphology was different from that in Test 2. In addition, Fig. 8.5b shows the change in maximum and minimum stress over time. The stress drop at the end of the life of Test 1 (pc-type) was slower than that of Test 2 (cp-type), suggesting that the cracks formed on the surface grew

Fig. 8.3 Morphology of longitudinal section near the center of specimen. **a** cp-type fatigue (Test 2). **b** pc-type fatigue (Test 1) [2, 15]



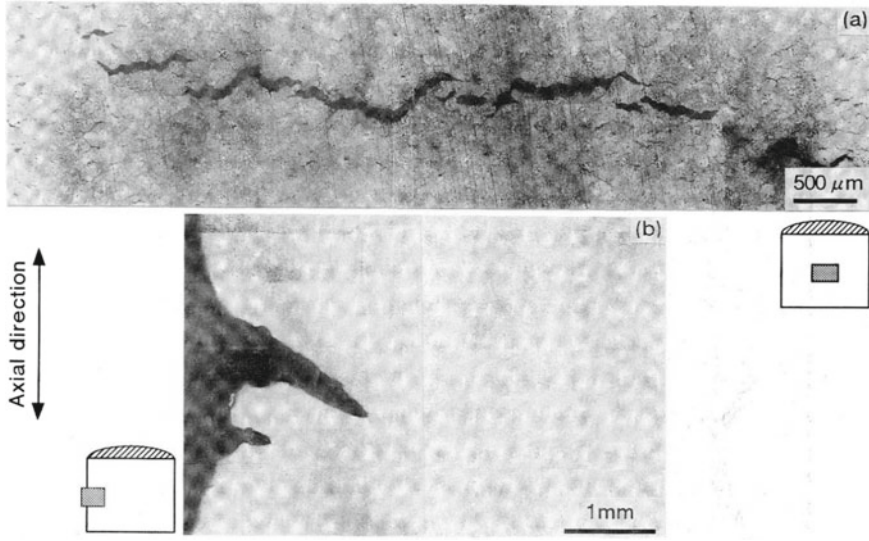


Fig. 8.4 Morphology of main crack on a longitudinal section of specimen. **a** cp-type fatigue (Test 2). **b** pc-type fatigue (Test 1) [2, 15]

stably. Furthermore, although it is not clear in Fig. 8.4b, the pc-type cracks grow inside the grains, so it can be said that they are not cracks formed by creep.

8.3.2 Extension of Life by Healing Small Inner Cracks

Figure 8.6 illustrates the failure lives of Tests 6 and 7 based on $(N_f)_{pc}$ and $(N_f)_{cp}$. In Test 6, pc-type creep-fatigue was performed after loading cp-type to $3/4$ of $(N_f)_{cp}$, and the life after changing to pc-type was 340 cycles ($=0.56(N_f)_{cp}$). The life predicted by the following linear cumulative damage law [10–12] is 151 cycles ($=0.25(N_f)_{pc}$), so the life is longer than the predicted life.

$$N_{cp}/(N_f)_{cp} + N_{pc}/(N_f)_{pc} = 1 \quad (8.1)$$

$(N_f)_{cp}$: Failure life of cp-type fatigue test

$(N_f)_{pc}$: Failure life of pc-type fatigue test

N_{cp} : Number of cycles of cp-type fatigue test

N_{pc} : Number of cycles of pc-type fatigue test.

Furthermore, when cp-type fatigue reaches $0.75(N_f)_{cp}$, many small inner cracks and cavities should have appeared as shown in Fig. 8.7 (Test 3), but when changing to pc-type, no small cracks or cavities were observed inside the damaged specimen. The main crack is a surface crack, as shown in Fig. 8.8, and is similar to that of

Fig. 8.5 Change in maximum stress and minimum stress. **a** cp-type fatigue (Test 2). **b** pc-type fatigue (Test 1) [2, 15]

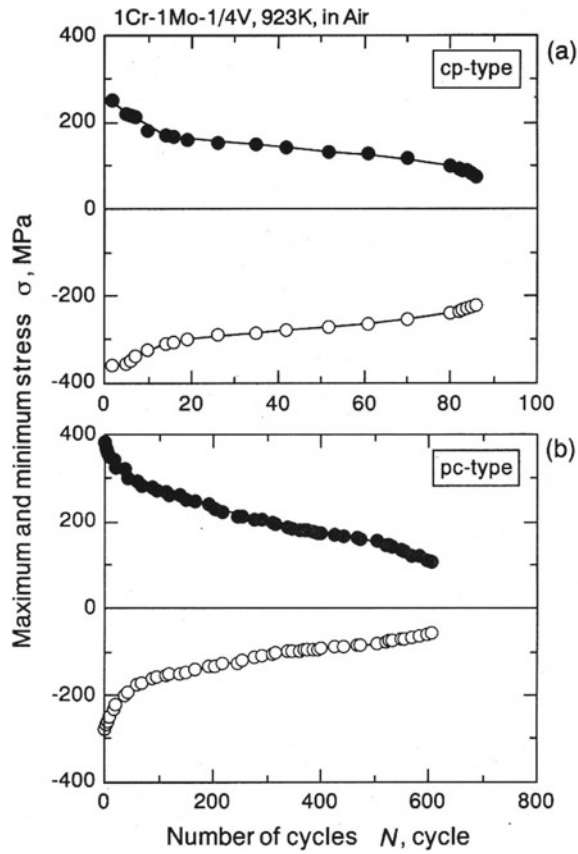
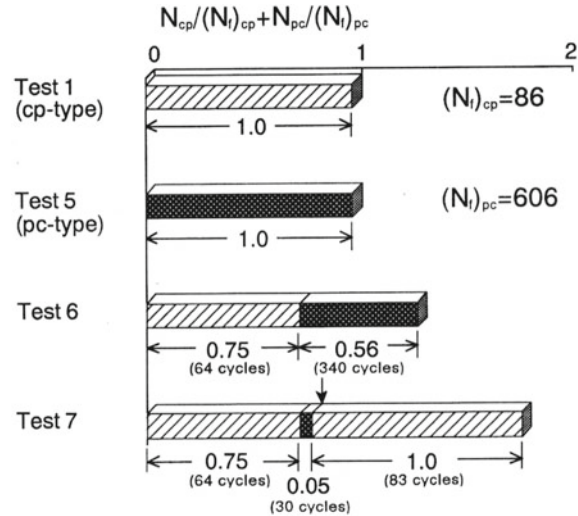


Fig. 8.6 Extension of cp-type fatigue life by subsequent pe-type fatigue [2, 15]



the pc-type (Fig. 8.4b, Test 1). Considering that the life of the pc-type was longer than predicted by the linear cumulative damage law, this suggests that the inner crack disappeared due to pc-type fatigue and was not involved in the main crack formation. The extension of creep-fatigue life due to the disappearance and healing of such small inner cracks and cavities supports that the small inner cracks are the main cause of failure of heat-resistant steel under high temperature, low tensile stress (or low tensile strain rate) conditions. Figure 8.9 shows the change in maximum and minimum stress over time in Test 6. The slow decrease in stress at the end of the life, similar to that in Test 2, suggests that the main crack propagated stably in Test 4, as in Test 1. However, the reason why the fracture life of Test 6 is shorter than that of Test 1 is that surface cracks have already initiated and grown due to cp-type fatigue before the waveform changes, and in pc-type fatigue, the surface cracks grow and break.

The results of Test 7 show that the specimen with $0.75(N_f)_{cp}$ damage due to cp-type fatigue was subjected to 30 cycles of pc-type fatigue and surface polishing was only applied halfway, and the subsequent cp-type creep-fatigue life was almost the same as that of virgin specimen (85 cycles). This can be interpreted as follows, referring to the results of Test 6. The inner cracks and cavities introduced by the initial cp-type creep-fatigue are largely annihilated by the subsequent pc-type loading. Surface cracks are also removed by surface polishing. In other words, at the point indicated by the arrow in Fig. 8.6 (at the start of cp-type loading in the latter stage), the specimen is almost in the same state as a virgin. Therefore, it is thought that the specimen was

Fig. 8.7 Morphology of longitudinal section near the center of specimen in Test 3 [2, 15]

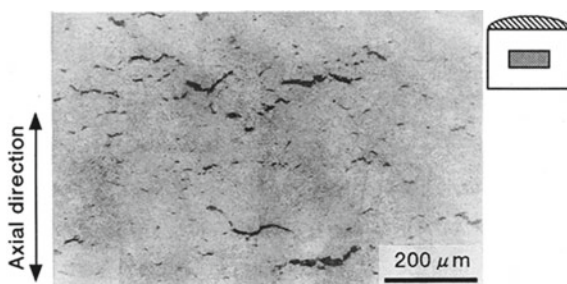


Fig. 8.8 Morphology of main crack observed in Test 6 [2, 15]

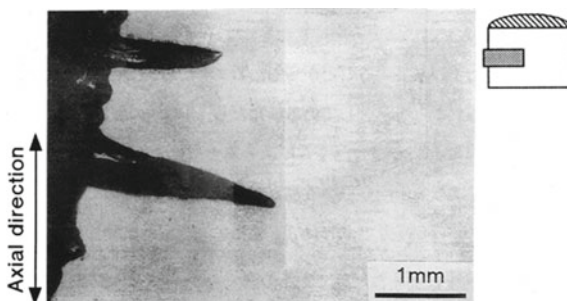
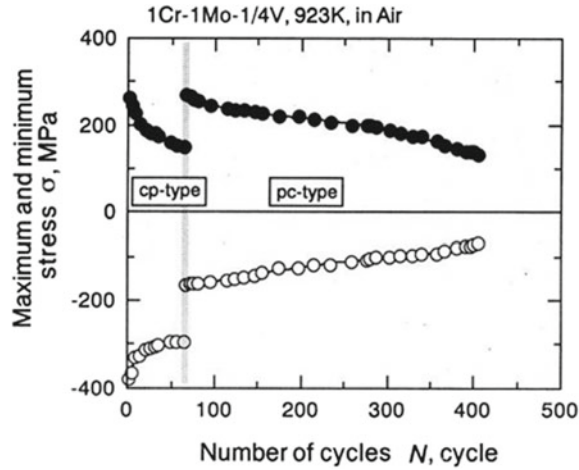
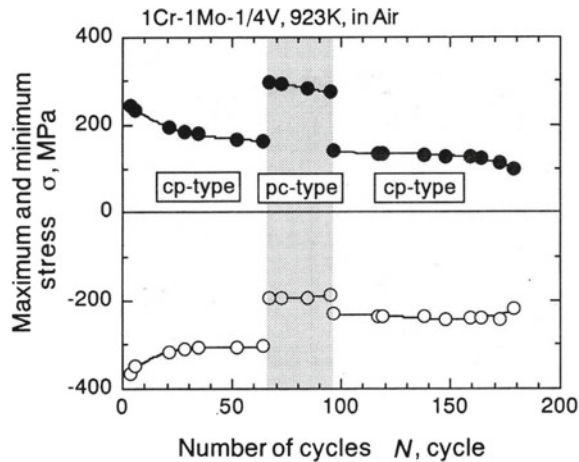


Fig. 8.9 Change in maximum stress and minimum stress in Test 6 [2, 15]



able to survive the same cycles as Test 2 in the latter stage of cp-type creep-fatigue. This suggests that internal damage formed by cp-type loading is almost completely healed by pc-type loading. In addition, Fig. 8.10 shows the change in maximum and minimum stress over time in Test 7. The stress change of cp-type in the latter stage is similar to that of Test 2, and no decrease in stress was observed until the end of the life, indicating that the main crack was formed at the end due to the coalescence of many cracks.

Fig. 8.10 Change in maximum stress and minimum stress in Test 7 [2, 15]



8.3.3 *Disappearance and Healing Process of Small Inner Cracks*

The results of Tests 6 and 7 showed that small inner cracks and cavities were annihilated and healed by pc-type loading. Therefore, we applied pc-type loading with different number of cycles to the specimen which introduced small inner cracks and cavities by applying 65 cycles ($0.75(N_f)_{cp}$) of pc-type loading (Tests 4, 5, and 6), the results of investigating the process by which small cracks and cavities gradually disappear are shown in Fig. 8.11a. For convenience, we refer to cavities with a half-length of less than $10\text{ }\mu\text{m}$, and cracks with a half-length of $10\text{ }\mu\text{m}$ or more. At the $0.75(N_f)_{cp}$ stage (Test 3), a large number of small cracks and cavities are observed at the internal grain boundaries. The crack density (the number of cracks with a half-length of $10\text{ }\mu\text{m}$ or more per unit area; the cavity density is defined in the same way) on the longitudinal section reaches 77 cracks/mm^2 . Furthermore, the cavity density with a half-length of $10\text{ }\mu\text{m}$ or less is 450 cavities/mm^2 . However, when pc-type fatigue is applied, the cracks rapidly become smaller and their number decreases. Figure 8.11b shows the changes in cavity density and crack density with increasing the pc-type load obtained from Tests 3, 4, 5, and 6. The figure shows the crack density distribution classified by length. From the same figure, it can be seen that by applying the pc-type load only five cycles, not only the cavity density drastically decreases to $180/\text{mm}^2$ and the crack density to $23/\text{mm}^2$, but also there is a clear tendency for the cracks to shrink. In other words, it can be seen that damage healing effects can be obtained even with a small number of pc-type loads. After applying pc-type creep-fatigue 30 cycles, small cracks were no longer observed and the cavity density was $59/\text{mm}^2$. In other words, healing has progressed further and the specimen is approaching the healthy state like a virgin material. This result corresponds to the pc-type fatigue result in Test 7, where the life after 30 cycles was comparable to that of the virgin material (Test 2). Figure 8.12 shows the crack length distribution for Tests 3 and 4. The vertical axis is the cumulative number of cracks with a length of $10\text{ }\mu\text{m}$ or more divided by the total number of cracks with a total length of $10\text{ }\mu\text{m}$ or more. Since the plots of Test 4 are located in the upper left corner compared to Test 3, the cracks observed in Test 4 are shorter overall, and all cracks are healed after a few cycles of pc-type loading.

8.3.4 *Healing Mechanism of Small Inner Cracks*

The following two mechanisms are thought to be at work for the disappearance of cavities and small cracks in the internal grain boundaries of the material due to pc-type loading.

- (1) Under high-temperature creep conditions, creep deformation near grain boundaries tends to become active [13], and compressive creep deformation near grain boundary cracks brings the upper and lower surfaces of the crack closer together.

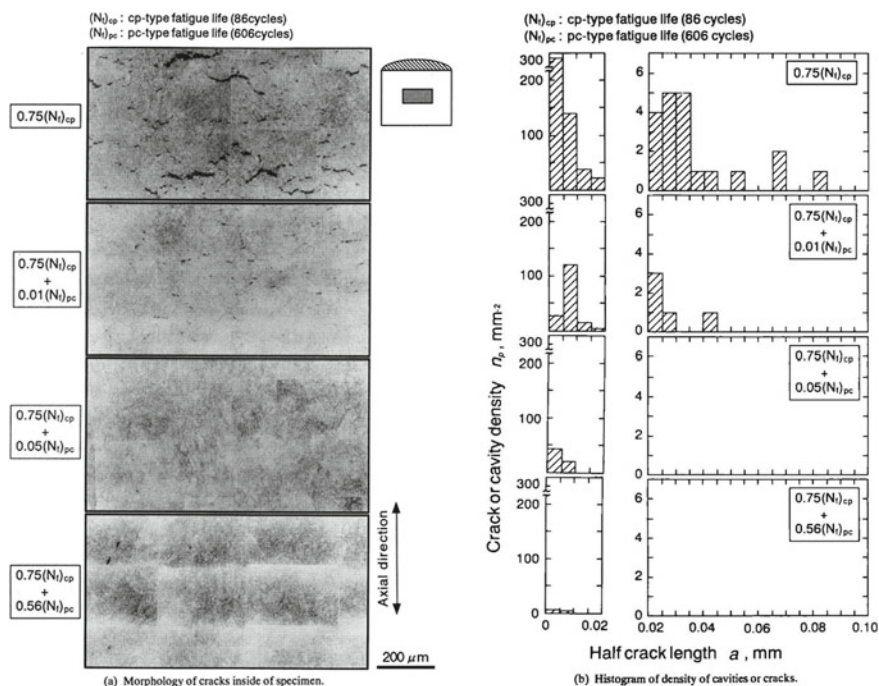
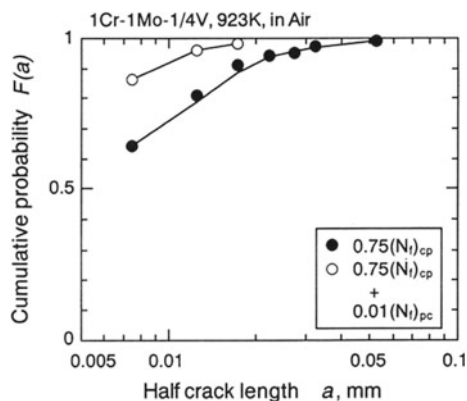


Fig. 8.11 Healing process of cavities and small cracks introduced during cp-type fatigue by successive pc-type fatigue [2, 15]

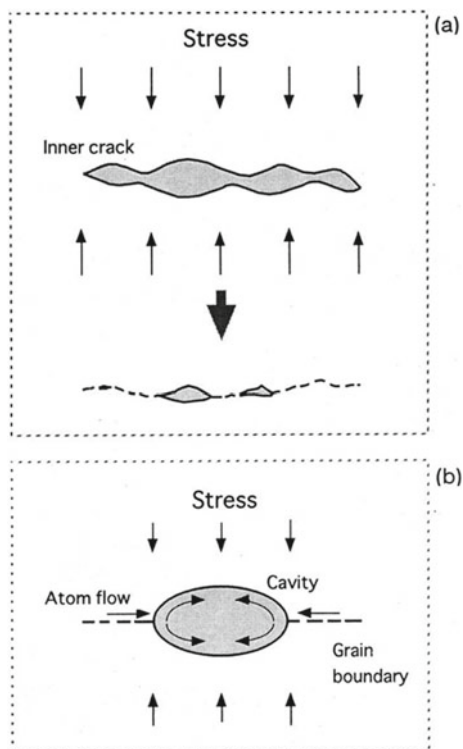
Fig. 8.12 Difference in distribution of inner crack length between Test 3 (●) and Test 4 (○) [2, 15]



On the other hand, the small cracks formed inside the specimen are under a high vacuum and the crack surfaces are very clean, so they easily coalesce. In other words, as shown in Fig. 8.13a, the crack splits and shrinks due to compressive creep deformation. Figure 8.14 is an example of crack healing.

Fig. 8.13 Schematic view of healing mechanism [2].

a Division and reduction of inner crack. **b** Reduction of cavity size due to grain boundary diffusion



- (2) Under compressive creep conditions, a flow (diffusion) of atoms occurs near the tip of the cavity as shown in Fig. 8.13b. This is a mechanism similar to the atomic flow that causes cavity growth in tensile creep [14], but because the flow direction is opposite, the cavity gradually shrinks. Figure 8.15 is an example of cavity healing.

Both mechanisms of (1) and (2) are due to compressive creep deformation, and it is thought that the compressive creep of pc-type fatigue caused the damage healing.

Furthermore, when the longitudinal section of Test 2 (Fig. 8.4a) was observed in detail, it was found that there were many small cracks and cavities near the tip of the main crack as shown in Fig. 8.16a, while no small cracks or cavities are observed above and below the main crack as shown in Fig. 8.16b. When a large crack forms, tensile stress is no longer applied above and below it, and small cracks tend to close. Therefore, it is thought that the cracks and cavities in this area disappeared due to the mechanism described above. This is one of the reasons why the crack and cavity density at failure is lower than that for $0.75(N_f)_{cp}$ in Fig. 4.4b in Chap. 4. In addition, the reason why cavities and small cracks are not observed in the internal grain boundaries of the material in the high-temperature, low-speed tension low-speed compression cc-type fatigue described in Chap. 3 is due to the suppression and healing of cavities and small cracks caused by compression creep.

Fig. 8.14 Morphology of healing of small crack. **a** Test 3 ($0.75(N_f)_{cp}$). **b** Test 4 ($0.75(N_f)_{cp} + 0.05(N_f)_{pc}$) [2, 15]

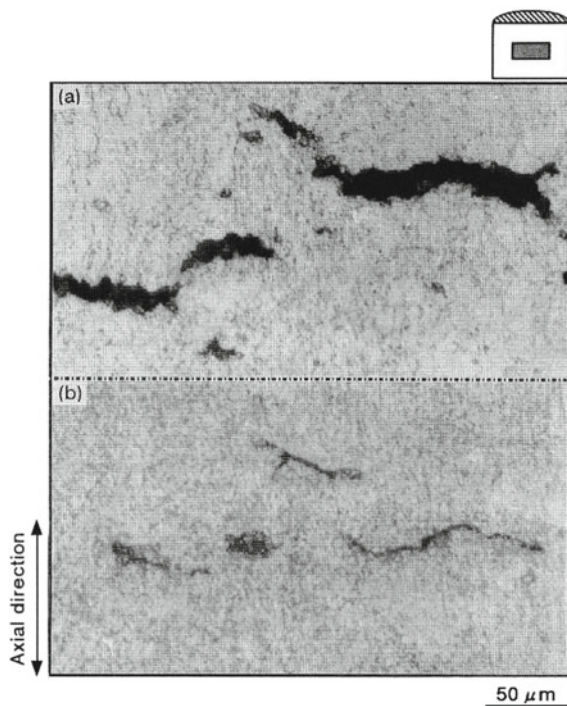


Fig. 8.15 Morphology of reduction of cavity size. **a** Test 3 ($0.75(N_f)_{cp}$). **b** Test 4 ($0.75(N_f)_{cp} + 0.05(N_f)_{pc}$) [2, 15]

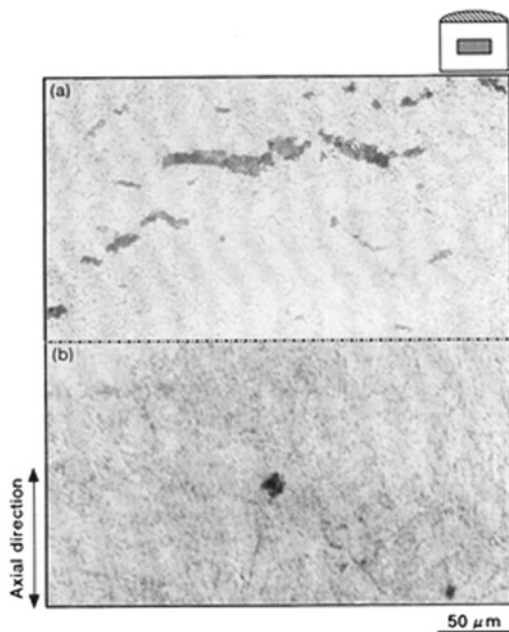
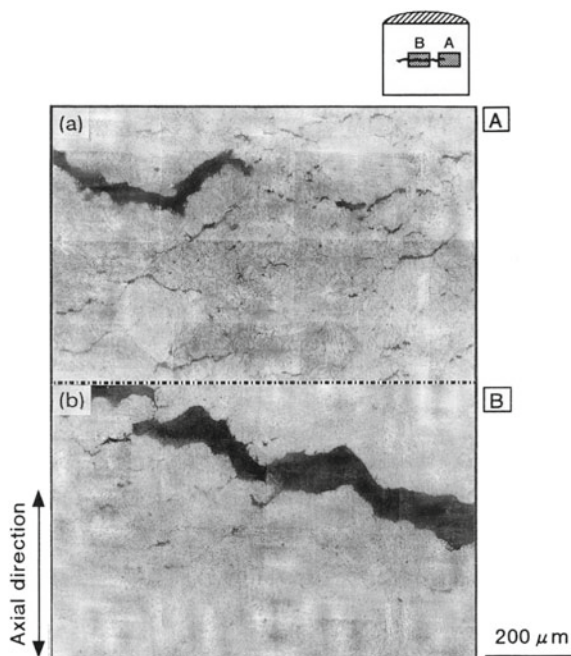


Fig. 8.16 Morphology of small cracks near the main crack observed in Test 2.

a Ahead of main crack tip.

b Near main crack [2, 15]



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Chapter 9

Conclusions



This book mainly describes the fracture behavior of heat-resistant austenitic stainless steel SUS304 under high-temperature creep-fatigue, especially under the conditions that exhibit inner cracking type fracture, from the onset to the final failure. Typical damage of austenitic stainless steel SUS304 caused by creep-fatigue is grain boundary cavities, small intergranular cracks, and large cracks. It is important for engineers to quantify these damages to predict the remaining life of high-temperature components and to improve the soundness of materials. As far as the authors know, a versatile damage evaluation method with clear physical meaning has not yet been established. Considering these backgrounds, we conducted creep-fatigue tests under various conditions and the creep-fatigue damage inside the materials was investigated in detail. Furthermore, a numerical model based on the obtained experimental results and the fracture mechanics was proposed, and simulations were conducted. In particular, we focused on appropriate quantification of materials damage caused by creep cavities based on statistical considerations, experimental observation and simulation of inner cracking type fracture, and healing of the damage caused by inner cracking by adding a strain waveform with reversed strain rates on the tension and compression sides. The main results are described below.

In Chap. 2, we statistically investigated the distribution of large number of creep cavities that were initiated on grain boundaries as creep-fatigue progressed and discussed the physical meanings of conventional damage parameters (cavity area fraction, areal cavity density, fraction of cavities on grain boundary lines, and A-parameter). We also evaluated these damage parameters for cavities that were actually observed. We believe that the obtained results are useful when considering which damage parameters are appropriate to use in creep-fatigue where the creep cavity dominates the damage.

In Chap. 3, we examined creep-fatigue conditions that indicate inner cracking type fracture. In general, creep-fatigue exhibits either surface cracking type fracture or inner cracking type fracture depending on the conditions. We conducted creep-fatigue tests varying the temperature, tensile strain rate, and compressive strain rate,

and clarified the conditions under which inner cracking type fracture occurs. Higher temperatures, lower tensile strain rates, and higher compressive strain rates tended to reveal inner cracking type fracture. In addition, we conducted interrupted tests under inner cracking type conditions and observed the longitudinal cross section of the specimens. Initiation and growth behavior of small cracks inside the material were clarified. Although the conditions for whether creep-fatigue reveals surface cracking type or inner cracking type depends on the material, the data obtained in this chapter will be useful for other materials as well.

In Chap. 4, creep-fatigue tests were conducted on austenitic stainless steel SUS304 and steam turbine rotor 1Cr–1Mo–1/4V forged steel. The initiation and growth behavior of small cracks inside the specimens was observed in detail. In both materials, cavities initiated on grain boundaries were coalesced with each other to form small cracks on one grain boundary, and continued to grow with the coalescence of the other small cracks in the later stage of life. Regarding creep-fatigue fracture of inner cracking type, it was shown that the initiation and growth behavior of cavities and small cracks is very important in evaluating the remaining life and integrity of materials.

In Chap. 5, we conducted numerical simulation of creep-fatigue fracture of austenitic stainless steel SUS304 under an inner cracking type condition. First, we explained the basic concept of a model of random fracture resistance of grain boundaries and then extended it to a three-dimensional model to perform a simulation for inner cracking type fracture. Since three-dimensional simulation requires three-dimensional grain boundary facets, a method of their generation was proposed. Although it is difficult to directly observe inner cracks experimentally, it has become possible to estimate their behavior based on the simulation results.

In Chap. 6, the above numerical simulation was employed to estimate the spatial distribution of small inner cracks. By selecting an appropriate crack initiation driving force F and crack propagation driving force K , the crack density, mean crack length, and crack length distribution actually measured on a longitudinal section can be well reproduced from early to late life.

Chapter 7 explained the influence of inner small cracks on large crack propagation behavior in austenitic stainless steel SUS304. Crack propagation tests were conducted in the following cases: when no inner crack existed, when inner cracks were initiated in advance, when the crack propagated coalescing inner small cracks initiated around the crack tip, and their mixed conditions. We clarified how much the crack propagation rate was accelerated compared to that when no inner crack exists in $dI/dN - \Delta J_c$ (creep J-integral range) or $dI/dN - \Delta J_f$ (fatigue J-integral range) relationship.

In Chap. 8, the creep-fatigue of steam turbine rotor 1Cr–1Mo–1/4V forged steel was examined. We experimentally verified whether cracks initiated under slow-fast type (cp-type or creep-plastic type) creep-fatigue can be recovered and healed by fast-slow type (pc-type or plastic-creep type) creep-fatigue. As a result, the cavities and small cracks that were initiated under the former (slow-fast type) creep-fatigue were recovered and healed by the latter (fast-slow type) creep-fatigue, and the service life was significantly increased. This result shows that creep cavities and small cracks

initiated inside the material under a high vacuum can be adhered and healed by applying a waveform that causes creep deformation on the compression side and is expected to be applied to the life extension of high-temperature components.

The results presented in this book are mainly those obtained approximately 30 years ago. In addition, since the target materials are limited to austenitic stainless steel SUS304 and steam turbine rotor 1Cr–1Mo–1/4V forged steel, it is necessary to carefully consider whether the same experimental method and simulation can be applied to other materials. In the coming era, it is expected that there will be a rapid increase in high-temperature components that deteriorate over time and highly accurate material damage assessment will be strongly required to ensure safe operations. We hope that you will be interested in the contents of this book and that you will find it useful.

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